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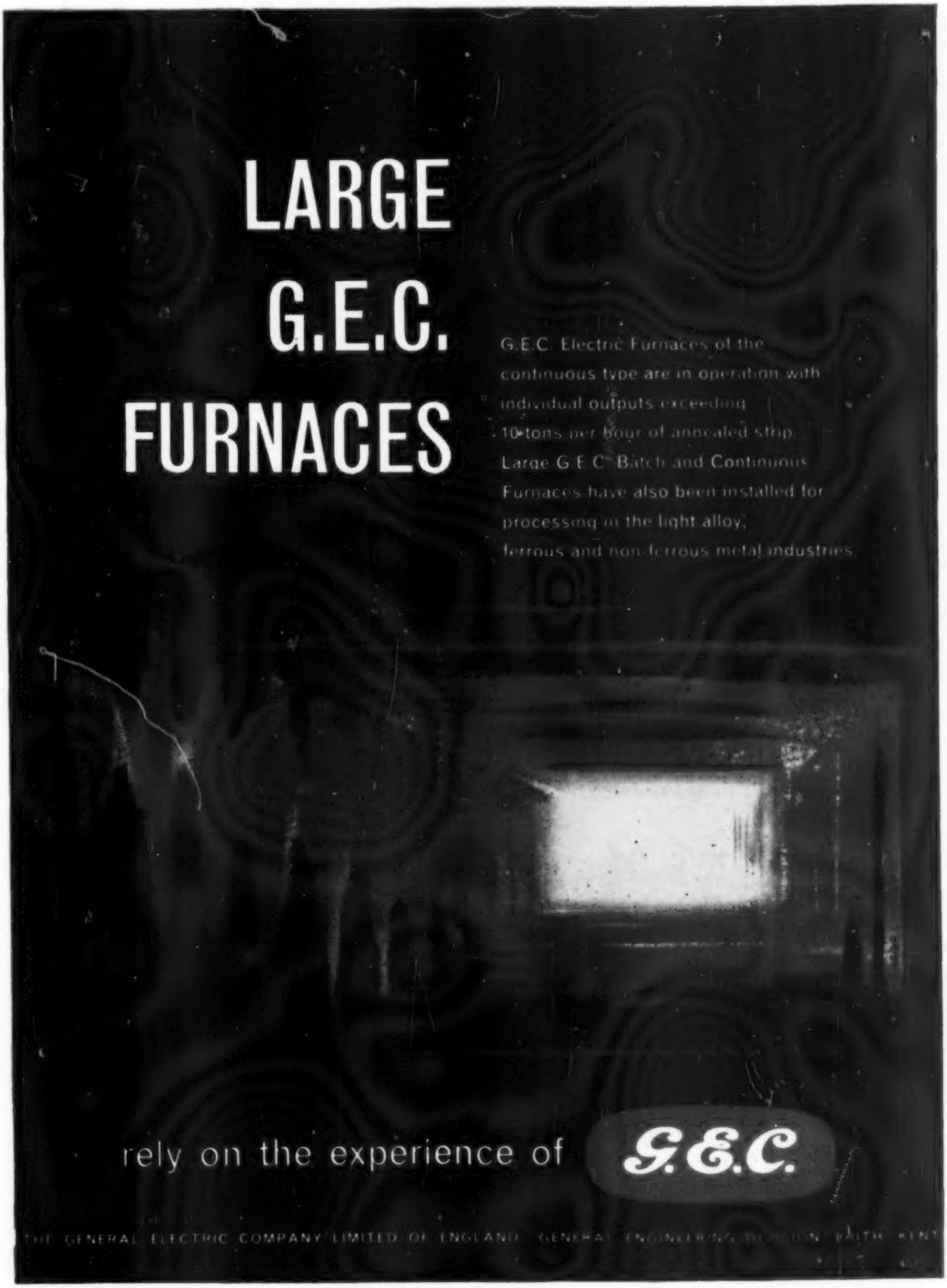
Vol. 28 : No. 185 FEBRUARY, 1961 Price 2/6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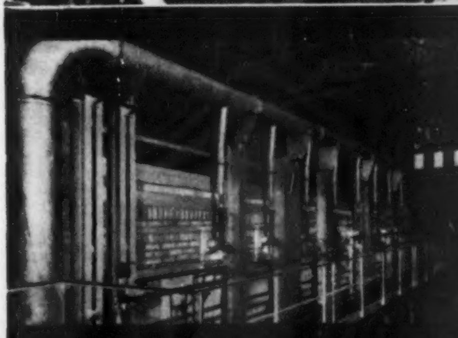

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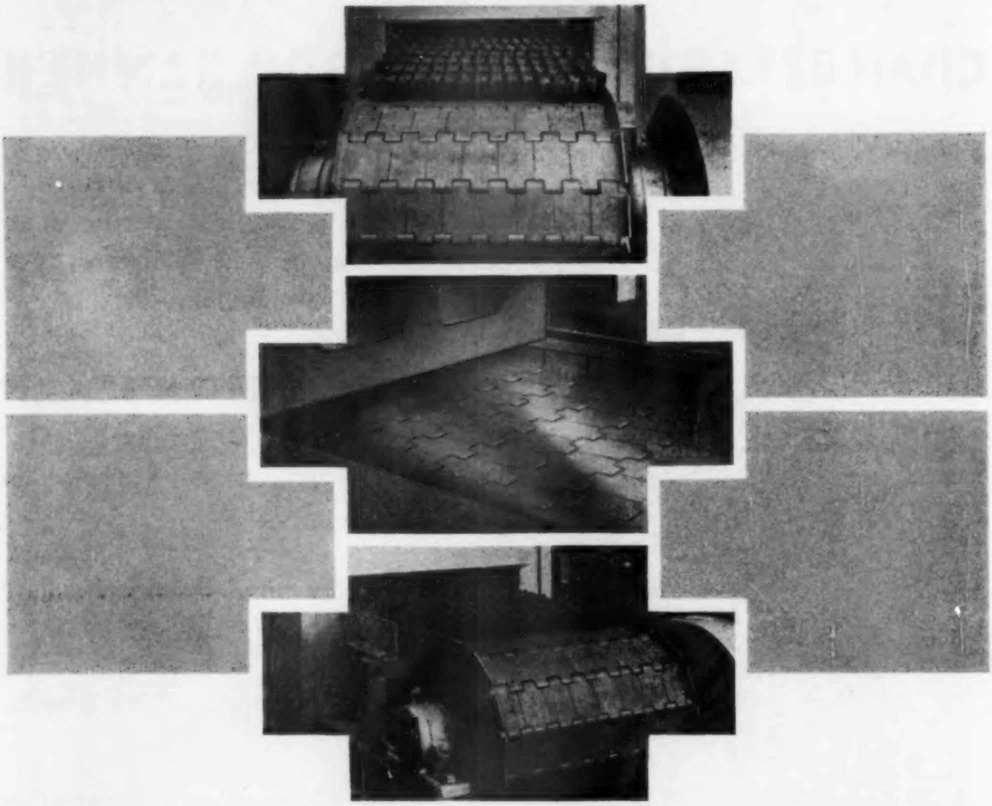

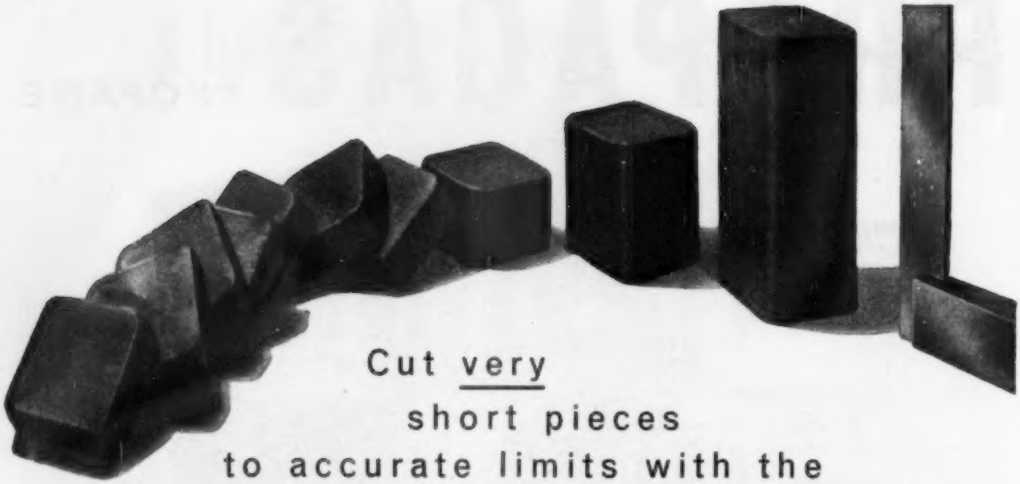

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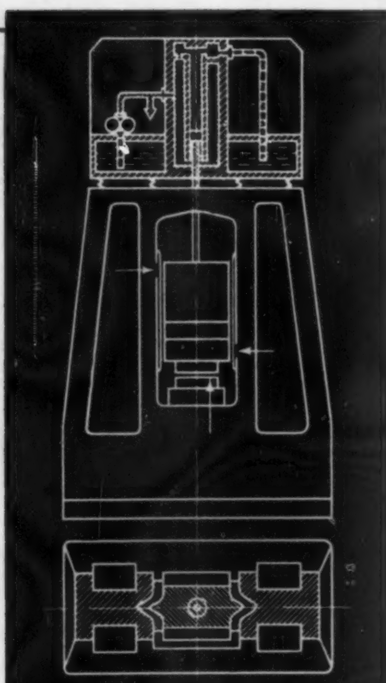

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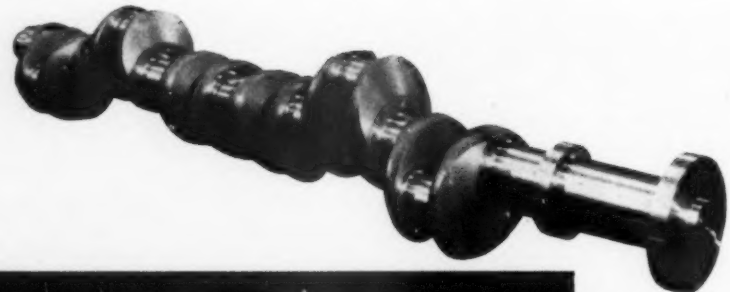

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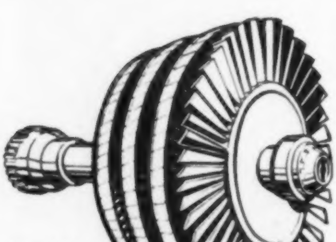

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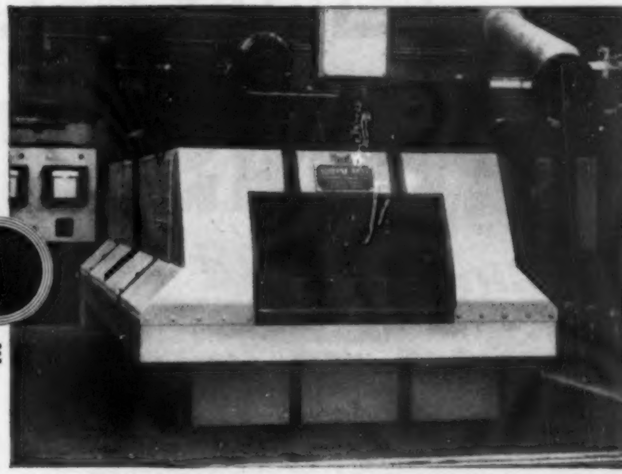

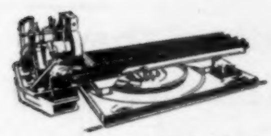

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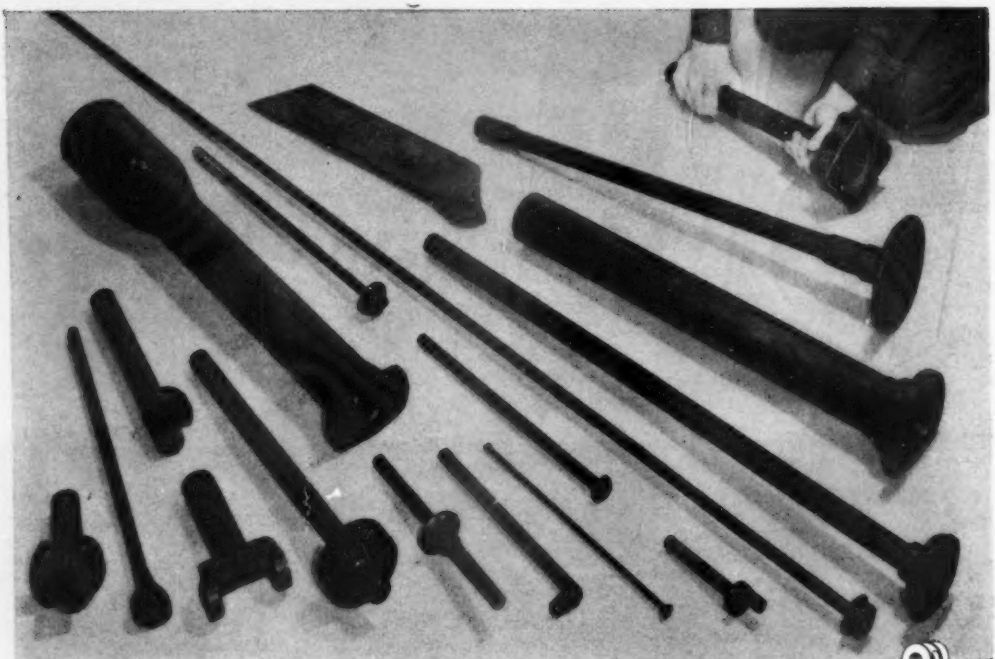


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February, 1961

Vol. 28, No. 185

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59 **Effect of carbide stringers on the distortion of die steels during heat treatment** K. SACHS, PH.D., M.SC., A.I.M.
The causes and mechanism of distortion of die steels during heat treatment, the influence of the structure of the steel and in particular the part played by carbide stringers are studied in this series of articles

63 **Precipitation hardening of low-carbon steel** MIRKO KLESNIL and PŘEMYSL RYŠ
Structural changes of a super-saturated solution of alpha iron in heat-treating experiments with a low-carbon steel were examined metallographically by optical and electron microscopy

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77 **Metallurgy in nuclear power technology** 6. Fuel element technology J. C. WRIGHT, B.SC., PH.D., A.I.M.
The metallurgy of nuclear power materials is developing on such a wide front and so rapidly that it is difficult for the non-specialist metallurgist to keep abreast with its scope. Dr. Wright outlines the subject in a series of articles

82 **News** 84, 86 **New plant**

Ergonomics

IT is a pity that the one remaining use of the classical languages today seems to lie in the endless coining of new names, thus emphasizing an already dangerous specialization. Whether the invented word is such a monstrous hybrid as automation, or justifiable derivatives as cybernetics or ergonomics, the result all too easily tends to produce a new cult of ' mumbo jumbo ' in the popular Press and a new esoteric specialized language. Perhaps this is unavoidable in our increasingly complex society but it does mean that useful information acquired in the process often remains locked up in a form not readily accessible to those responsible for policy decisions in industry.

Ergonomics is the scientific approach to the physical factors of human performance. The Americans have the term 'human engineering' which although ambiguous has the advantage of stressing the important elements. Ergonomics is essentially the study of man in relation to the physical factors in the environment in which ' work ' is done—work being expenditure of energy in the physiological sense. Stated in this form it may seem to be a study not very closely related to the everyday needs of industry, but this is far from being the case. It is in fact a good example of useful information locked up in a specialized language, and it offers results which are by no means being as widely employed in industry as they deserve.

An attempt was made to show the scope of the study of ergonomics to the world of industry by Dr. C. B. Frisby, PH.D., director, National Institute of Industrial Psychology, at the Royal Society of Arts, last month. In the course of his lecture, Dr. Frisby grouped the physical factors in the working environment as follows: 1 *The surroundings*. Lighting, heating and ventilation. Noise and vibration. Dust, odours and toxic substances. 2 *Tools and equipment*. Hand tools, jigs, benches, seating. Equipment used for lifting and transporting material. 3 *Plant and machinery*. In particular, the nature and location of displays, which indicate to a man the way the machine is performing, when action is required on his part, and the results of this action. Secondly, the nature and position of controls used for operating the machine or system.

Speaking of plant and machinery, Dr. Frisby said that the designer had been faced with mechanical problems of increasing complexity as machines had been demanded to undertake more and more operations formerly recognized as skilled. Perhaps this was one reason why so much industrial machinery of the highest mechanical ingenuity appeared to have been designed with very little thought for the capacities and limitations of the human beings who were going to be responsible for its operation. He cited the example of power presses used in the production of countless small metal stampings. Very many moderate-sized power presses imposed on the operator a posture of extreme discomfort, seated in a crouching position with arms thrust forward and outward, one leg bent under him operating the treadle controlling the descent of the head of the press.

Why was it, asked Dr. Frisby, that in the industrial field there was still so much that was deplorable in its human engineering? Weston in 1923 advanced the theory that machines were badly designed because the designers did not use them, and thus were not led to recognize their failings from the operator's point of view. This had always seemed to him to be a primary reason, but he was beginning to wonder.

In the long run, he supposed it was the attitude of him who paid for a machine which had the greatest influence on the designer. If the managers of industrial undertakings became more aware of the handicaps imposed on their workers, handicaps which were reflected in loss of output, in sickness absence and labour turnover, they would perhaps demand that machine manufacturers should take more account of human capacities and limitations in the design of their equipment.

Russian forging journal

Abstracts from the Russian forging journal — Kuznechno - Shtampovochnoe Proizvodstvo, July, 1960, **2.** *This is the second year of this journal devoted specifically to forging. We shall try to give indications of the contents of future numbers in* METAL TREATMENT *each month.*

Explosion working of metals. V. G. KONENKO. Pp. 1–4.
The method is characterized by the high energy capacity per unit volume, the high velocity of movement of the deforming tool and proportional decrease in its weight, and consequent lightness, portability and cheapness of the working equipment by comparison with conventional stamping and forging machines.

Testing sheet metal at high rates of deformation. Pp. 5–8.

The forging of heat-resistant steel ingots. F. M. VALYAVKIN. Pp. 8–13.
Difficulties in forging high-alloy, stainless, heat-resisting and other alloys are discussed and a report is given on research to determine the optimum forging technique for 23Cr–18Ni steel ingots. A study was made of the plastic properties in the hot state, the forging temperature ranges were determined, and control of the ingot heating conditions was examined. The design of tools and the technical forging process were studied. The effect of these various factors on the quality of forgings is assessed.

The process of forming cutting edges (on agricultural implements) by plastic deformation. S. SH. YASHAYEV. Pp. 13–15.
In response to readers' requests further details are given of the process described by Noritsyn and Yashayev in Kuznechno-Shtampovochnoe Proizvodstvo, 1960, No. 1.

A new technical process for stamping piston rods. G. S. ZAIKOVSKII, M. I. DYTYNENKO and I. A. FOMENKO. Pp. 15–17.

Investigations of the temperature and rate of working as factors in the plastic deformation of metals. N. P. AGEEV. Pp. 17–21.
A description is given of sundry test machines to investigate these factors.

An experimental investigation of a hydraulic screw press hammer. YU. A. BOCHAROV. Pp. 22–27.
Some technical characteristics of the press are: nominal force 25,000 kg., maximum pressure of the hydraulic fluid (oil) 200 kg./cm.^{2}, producing a maximum static force of the hydraulic cylinder of 5,000 kg.; maximum travel 210 mm. and working stroke 15–39 mm.; nominal power of drive motor 3·8 kVA. with nominal r.p.m. 970. The designs of the press have already been given in Kuznechno-Shtampovochnoe Proizvodstvo, 1959, No. 8, and the present article describes research carried out with the first experimental press.

A hydraulic damper for crank-drive drawing presses. YU. M. BUZIKOV. Pp. 27–31.
The results are given of testing this device to protect the press and forge stock from harmful impact shocks. Use of such a damper permits a considerable increase in the speed of running of the press and a substantial increase in the productivity of shops equipped with such presses.

Plastic deformations of die base plates of hydraulic presses. G. Z. ZAITSEV. Pp. 31–36.
The article presents the results of an investigation of certain laws governing the accumulation of residual deformations and methods of increasing resistance to these processes applicable to such plates in power presses. Comparative fatigue tests were conducted on hardened and unhardened 5Cr-Ni-V steel and medium-carbon steel specimens at normal and elevated temperatures by flexure of a fixed specimen on a rotating load-fatigue testing machine. The effect of the resultant residual stresses is preserved at higher temperatures, falling somewhat on heating to 450°C.

Increasing the life of piston rods of hammers by burnishing with rollers. M. YA. BELKIN. Pp. 36–38.
Increased resistance to fatigue fracture in such components is connected with the creation of residual compressive stresses in the surface layers of work-hardened parts, thus compensating the high-tensile stresses in service. Their fatigue limit in service may be greatly improved by as much as six times by this method of work hardening.

A new 2,000-ton multi-stage press. V. I. KHABAROV and S. M. NEKHAI. Pp. 39–42.
The new press can exert a specific pressure up to 130 kg./cm.^{2} on components up to 1,500 × 1,000 mm., and has 11 stages. Hydraulic operation is automatic. The press is described in detail.

Hammer heads for working tungsten and molybdenum on rotary forging machines. B. S. SEKUN. Pp. 42–43.
The article outlines the use of tungsten and cobalt carbide powdered carbides for such heads and their method of manufacture.

Preheating of roll dies. L. N. ZHUBR and N. A. NOGA. P. 43.
A gas torch is described for heating the rotating dies on forging rolls.

Experience in embossing designs on table cutlery. A. D. ROVNII and I. K. GROM. Pp. 45–46.
Fractures of embossing dies are analysed.

The forging of titanium

A symposium on titanium production methods in the aircraft industry was held by I.C.I. Metals Division last year at the I.C.I. Titanium Plant, Waunarlwydd. Three of the twelve contributions were specifically devoted to titanium forging and are given below by courtesy of I.C.I. Ltd.

Drop-forging titanium alloys

J. C. BELL, A.R.Ae.S., and E. LEYBOURNE, A.I.M., A.MET., *Firth-Derihon Stampings Ltd.*

FIRTH-DERIHON have had considerable experience since the early days of titanium in producing different types of component, either as hand forging or drop forgings, for the aircraft industry. This metal, with its high strength/weight ratio at room and reasonably high temperatures, and its corrosion resistance, offers the designer many advantages, and many components are now being called for in titanium and its alloys.

The production of drop-forged compressor discs has been chosen for brief discussion.

Equipment

Allowing for the fact that resistance to deformation is greater than that of typical alloy steels, it has been found from experience that the use of heavier equipment is advantageous, although material displacement must be controlled. Where one can be selective, something in the region of a 20% heavier-type drop stamp assists considerably in easing deformation, without giving unduly heavy blows.

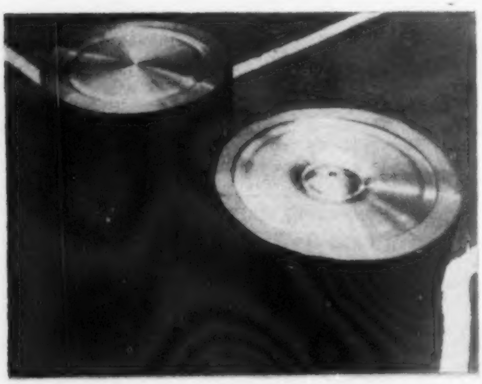
Town's gas-fired furnaces are used, automatically controlled both for atmosphere and temperature. Cleanliness of the chamber, especially the hearth, is essential; if necessary, the bricks are cleansed to ensure the most satisfactory conditions. With this in mind, a refractory platform of clean fire bricks is essential to keep the billets clear of any possibly contaminated zone; as the pick-up of impurities is rapid, any ferrous scale which may be left in the furnace can cause rapid oxidation of the titanium.

Dies and tools

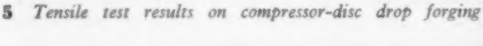
Die blocks in steels to standard specifications, such as nickel-chrome-molybdenum to B.S.S. 224, No. 5, with a Brinell hardness of 302–277 (3·5–3·65) and cut in the oil-hardened and tempered condition, have been found generally suitable for larger forgings.

When new dies are cut specially for a titanium forging, it is desirable to allow proportionate die-block sizes larger than when producing steel drop forgings of similar form, bearing in mind the stiffness of the material, the restricted range of

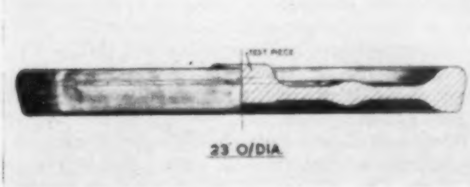
1 *Various components in ICI* 230, 314, *and* 317



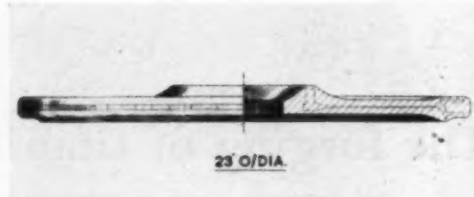
2 *Dies for compressor disc forging*



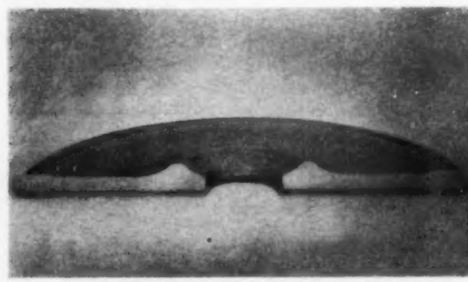
5 *Tensile test results on compressor-disc drop forging*



6 *Compressor-disc drop forging* 23 *in. o.d., rim thickness* $2\frac{1}{8}$ *in., diaphragm thickness* $\frac{5}{8}$ *in.*



3 *Compressor disc drop forging* 23 *in. o.d., rim thickness* 1 *in., diaphragm thickness* $\frac{5}{8}$ *in.*



4 *Section of compressor disc forging*

7 *Smaller discs in titanium* 679 *(as-forged condition)*

forging temperatures involved and the greater force necessary for material displacement.

In some cases, dies previously used for steel drop forgings can be used for producing titanium forgings, provided that the surface finish of these dies is smooth and clean. However, this is not always advantageous, as the general contraction allowances for titanium are less than those for steel.

Impressions fill up smoothly, even though the complex forms on the larger components may sometimes result in the material moving slowly. It is important, therefore, that the finish on the dies and tools is smooth, and even on disc dies a polishing operation is advantageous. Any tool or die sinking marks are reproduced on the relevant impression faces, and could be detrimental to the forging if not controlled. Where possible, the easing of radii, together with more than standard taper as necessary, also assist the movement of the material and reduce the chances of sticking. These precautions have a direct effect on improved die life, which is usually less than that for dies used for alloy steel drop forgings.

Forging and stamping

Fig. 3 shows a compressor disc drop forging produced by the author's company. The outside diameter is approximately 23 in., with a rim thickness of 1 in. and a diaphragm thickness of $\frac{5}{8}$ in. The material is I.C.I. titanium alloy 679, which can be heat treated to give high-tensile and creep figures of 400–500°C., of which more details are given later.

One of these particular discs has been sectioned (fig. 4) to show the general form. Radii have been formed to ease material flow and the only taper for draw in this instance is from the outside diameter radii to the line of the flash. An allowance for an integral test piece has been made in the centre, adjacent to the punched-out portion.

As regards upending ratios for discs, there is no necessity—unless particularly requested by the customer—to work to more than 1/1 as even at these proportions, allowing for the forms involved, there is a considerable amount of material displacement which produces satisfactory flow in the required directions.

Pre-heating of the machined billet is carried out at approximately 700°C., the piece then being brought up to 920°C. in preparation for forging. It is most important that the transfer from the furnace to the forging unit is made as quickly as possible, to permit forging over the small temperature range necessary for obtaining the most suitable physical properties.

Before drop forging in the finishing dies, the billets are knocked down into cheeses during the forging operation, after which they are slow-cooled, preferably in a suitable furnace. Pre-heating is then carried out as necessary, prior to heating to forging temperatures for final drop forging. The number of heats is kept to a minimum, bearing in mind physical properties and the need to limit the chance of contamination.

Usually quite a number of heats is necessary to produce a satisfactory drop forging within the limits requested by the customer but, in general, drop forgings of proportions similar to those on the diagram can be considered as practical for production, even allowing for increased diameters which could be classed as within the limits of plant available.

During drop forging, some forgings require punching and trimming at intermediate stages, before finally punching and trimming after the last heat. The hot trimming, or clipping and punching, is clean and is not considered likely to be troublesome. A sand or light shot-blast finish is usually given before final inspection prior to despatch.

At all stages of production there is need for supervision by a competent staff of technicians. Full laboratory and metallurgical control is also essential throughout the manufacturing cycle.

Tests

The disc shown was heat treated as follows:

$\frac{1}{2}$ h. at 900°C., air cooled
24 h. at 500°C., „ „

The following results were then obtained on tensile tests (fig. 5):

Tests at the rim and diaphragm positions—70 ton/sq. in.

Centre boss and internal positions—68 ton/sq. in.

General

Fig. 6 shows another disc with a diameter of approximately 23 in. thickness at the rim being $2\frac{1}{8}$ in. and at the diaphragm portion $\frac{5}{8}$ in. In this instance, an allowance has been made at the centre portion for an integral test piece. The smooth-flowing radii can again be seen; these assist in easing the material displacement.

Fig. 7 shows a further batch of discs in titanium 679, in the as-forged condition. These discs are smaller than those generally referred to earlier.

Temperature control is most important at all stages of production, and is essential to ensure a satisfactory finished component. In general, taking into consideration the lower working temperatures, the movement of this material can be considered as not unlike S. 62 type stainless steel. The forgings are, in the main, easier to handle, because of the lower density of the material as compared with alloy steel.

Press-forged titanium components

E. T. STEWART-JONES, *High Duty Alloys Ltd.*

High Duty Alloys has handled a certain amount of titanium from about 1952, and during the past three or four years has made over 20,000 miscellaneous forgings and around a quarter-million compressor blades. These forgings have ranged in weight from 800 lb. to a few ounces and, although not all have been made from I.C.I. material, a large proportion of the forging stock has originated at Witton.

Quality of titanium

Over the past few years the company has examined metallurgically at least 200 forgings, besides samples representing every batch of stock from which forgings have been made. From this experience, it can be stated without reservation that the quality of titanium has been found to be consistently good. This satisfactory experience reflects the effort which has been lavished on titanium by ICI and others, and it is this sound technical foundation which has assured for titanium an expanding place amongst engineering materials. It is interesting to note that high-tensile steel, which has begun to share with titanium the benefits of vacuum melting and casting techniques, has now become a much cleaner material, with notably higher transverse ductility, lower brittle/ductile transition temperatures, etc.

Press-forging technique

This paper is particularly concerned with press forging. There are two ways of forging titanium. One is to strike it hard enough and with sufficient frequency to maintain its temperature within the rather limited range in which it is plastic enough to move, but without overheating; the other is to press or squeeze it at a controlled rate, again so that by the dissipation of energy the temperature of the mass is kept within the plastic range.

The word ' squeeze ' has been used to suggest the action of a hydraulic press, but titanium forgings are made also on mechanical and screw-type presses. These give in effect a quick dab, in contrast to the hydraulic presses which can squeeze either quickly or slowly according to whether fed by direct-acting pumps or accumulators. Mechanical and screw presses and direct-pumped quick-acting hydraulic presses range in capacity from around 100 tons up to about 2,000 tons, while the accumulator-fed hydraulic presses range up to 12,000 tons in this country and up to 45,000 tons in America.

It is interesting to consider the potentialities of a 12,000-ton press as a tool for forging titanium, because this is a relatively unexploited field, and we must attempt first to define the sort of forgings that such a press could make.

If a forging can be tackled piecemeal—like a long spar, or a rough oversize shape without much form —then only a small press is needed to make a large forging. At Witton, for example, a hydraulic press of only 1,500-ton capacity is used to cog down cast ingots weighing up to 2 tons, and by the same token a big press could make a relatively huge forging.

When considering a die forging, however, made with a single stroke in closed dies to close dimensional limits (and particularly when the section thicknesses are small), then even a press of 12,000-ton capacity can make only a moderate-sized forging in titanium. To be more explicit—talking in terms of aluminium and, with some reservations, steel—we find that a specific pressure of about 10 ton/sq. in. is needed to make a simple die pressing and a pressure of perhaps 20 ton/sq. in. for a difficult one. With 12,000 tons available, we can make a simple die forging of plan area up to 1,200 sq. in., or a difficult configuration of up to 600 sq. in. With titanium, it may be necessary to use nearly double the specific pressure, which gives a size limitation for die forgings ranging in plan area from 600–300 sq. in. for the simple to difficult types.

These limits have been expressed in very conservative terms to emphasise the difference between ' close tolerance ' forgings and those on which there may be generous machining allowances. For example, compressor disc die pressings have been made up to 3 ft. dia., with a plan area of 1,000 sq. in. It must also be noted that we are concerned only with the plan area of the material actually under pressure, which may be much less than the overall area of the component when account is taken of cut-outs and other enclosed voids.

Generally speaking, thick, bulky, ' oversize ' die forgings are easy whilst wide, thin, ' close tolerance ' die forgings are difficult. But account must be

taken of the means whereby a difficult forging can be made easier, for example, by using a series of dies (so-called 'blockers' and moulding dies) by which the final shape is arrived at in easy stages. There is also hope that the future development of alloys with higher permissible forging temperatures will give titanium the same high plasticity as steel.

Machining is always costly, and it may be handsomely rewarding to make a pair or possibly a series of press dies to produce a press-forged component on which relatively little machining will be needed, instead of a rough-forged shape which is also extravagant in terms of raw material. Another advantage of the press-forging technique is that it allows the possibility of throwing up vertical (or draftless) ribs or flanges, again avoiding unnecessary machining time and metal loss.

A considerable part of the cost of a die forging is in the cost of the dies, both in terms of the die material and of the time taken to machine an impression. Fortunately, the relatively gentle action of a hydraulic press allows the use of comparatively small shallow die blocks, and the risk of premature failure by breakage in service is slight.

Titanium alloys

Forgings have been classified into easy or difficult types, and the same distinction must be made between the various titanium alloys, some of which, like ICI 314A, are admirable for making accurate press-forged components, within the 65-ton range, to close dimensional limits. The newer, more highly-alloyed mixtures, like EX 013C, are distinctly more difficult, and discourage the forger from making rash promises about holding close dimensions.

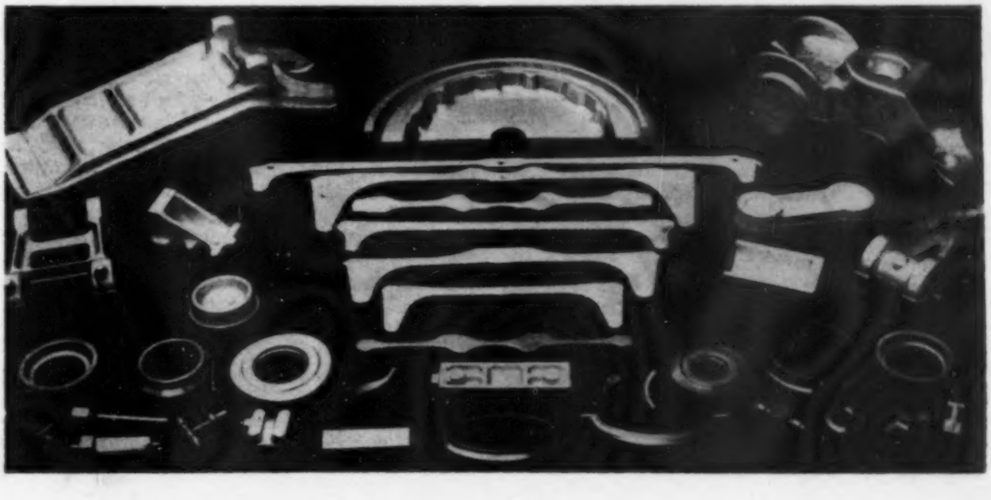
Titanium is in practice not nearly so temperamental as was at one time feared. It seems characteristic of all alloys to give admirably consistent tensile properties in cut-up forgings, with notably high transverse ductility values. Neither hydrogen pick-up during processing nor oxygen contamination have proved to be serious, and the surfaces of forged components on which there will be no machining before service can be effectively cleaned by grit blasting and acid pickling. Acid pickling can also be used in certain circumstances to control dimensional limits, the titanium being dissolved so uniformly as to leave an acceptably smooth surface finish.

Heat treatment, whether involving annealing or quenching, presents no difficulty, though the weakness of titanium at the higher heat-treatment temperatures may necessitate the use of some form of jigging to prevent sagging.

Ring rolling

There are now in this country a number of installations well suited for handling titanium by ring rolling. The company's own ring roller will roll up to 6 ft. dia. and a depth of 10 in., and when its full potentialities have been proved it is expected to be rolling plain and shaped rings in all titanium alloys.

8 *Miscellaneous die forgings in titanium, including large press-forged components and compressor discs ranging from 17 to 36 in. dia. with ½ in. diaphragm thickness*



Titanium compressor-blade forgings

G. O. ECCLES, A.M.I.MECH.E., *Rolls-Royce Ltd.*

This paper is a short historical account of the experience of Rolls-Royce in manufacturing compressor blade forgings in titanium and some of its alloys. Before dealing with titanium blades in particular, and for the benefit of those who are not familiar with these engine components, a brief explanation of the general principles of manufacture might be of value.

The working part of a compressor blade consists of a thin aerofoil section. At one or both ends is an enlarged piece for attaching the blade to the appropriate engine components. The majority of blades, and particularly those in titanium, have this feature at one end only and, as far as the forging is concerned, it is often of roughly cubic shape. It is generally called the root fixing; the side of the cube to which the blade is attached is called the root platform. The ratio of the area of this platform to that of the adjacent blade section can vary from about 3 : 1 to as much as 30 : 1. This has a considerable influence on the cost of manufacture, particularly if there is a platform at both ends of the blade. It also largely dictates the method of making the forging and the amount of tooling required.

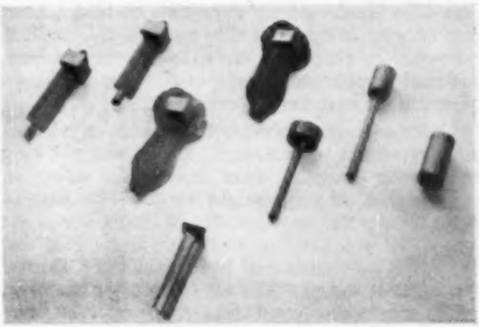
Before attempting to make a blade it is necessary, in the interests of material economy and forging-die life, to distribute the raw material so that its proportions in respect of volumes and cross-sectional areas are similar to, but a little greater than, those of the finished blade. The forging, called a dummy at this stage, is in its simplest form form of circular cross-section. This type of dummy is also the cheapest to manufacture, as the tools required are relatively simple and can be discarded as soon as wear becomes troublesome. It is produced by hot-working operations, such as mechanical or electrical upsetting, from small-diameter bar, or by the extrusion of slugs or billets, cut from bars of larger diameter. In the latter method, now standard practice, it is usual to limit the extrusion ratio to about 8 : 1, so a further upsetting operation on the large end is usually necessary. At this stage, the surface of the dummy is cleaned up either by emery dressing or by centreless grinding, and it is then etched and inspected for defects which must be removed before forging. Figs. 9–12 illustrate stages of manufacture.

Titanium compressor blades

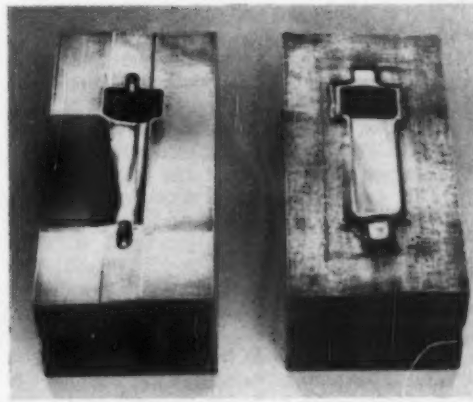
Our first experience with titanium was in 1950, when a small sample of T 150A material was obtained from America. This was an alloy containing about 1·5% iron and 2·5% chromium. From this sample, a few Avon compressor blades were made, and subsequent engine performance was sufficiently good to encourage the designers to make a much wider use of the material in later designs of engine.

In 1951, about 400 lb. of small-diameter titanium bar was received from the Titanium Metals Corporation of America and also the first sample batch of hot-rolled T 75 material produced in this country by ICI. These materials were used for eight rows of blades in the high-pressure compressor of the prototype Conway engines. The iron-chromium alloy was used for the rotor blades and the commercially pure material for the stator blades.

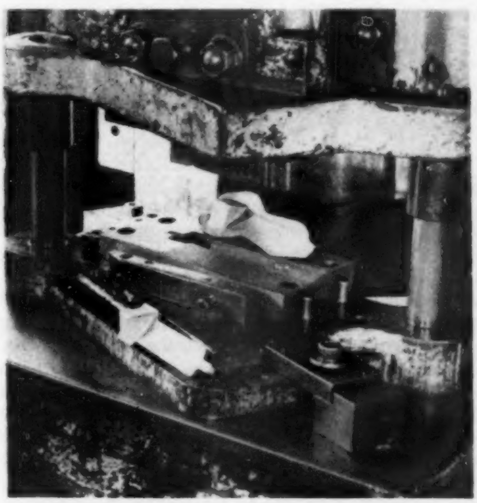
Not unexpectedly, a number of problems arose with both materials. For instance, the American material had been heavily cold worked and a number of pieces split from end to end on being dropped on the floor. Subsequently, coarse grain and cracking troubles were experienced, probably a result of cold working the bar material. On the other hand the commercially pure material, also of small diameter, caused trouble because of rolling defects and internal faults associated with the manufacture of the ingots. The dummies for these blades were made by the electrical upsetting method and were forged into blades in one operation, this being dictated by the relatively small size of the blades in question. Another feature of these blades was their maximum thickness, which at 50 mils was considerably less than the sample Avon blades previously mentioned, so that they were inherently more difficult to forge. Furthermore, the information available at that time was that a forging temperature of 800°C. should not be exceeded, and these facts, together with difficulties of lubrication, resulted in a considerable amount of die trouble. Severe abrasion and local welding



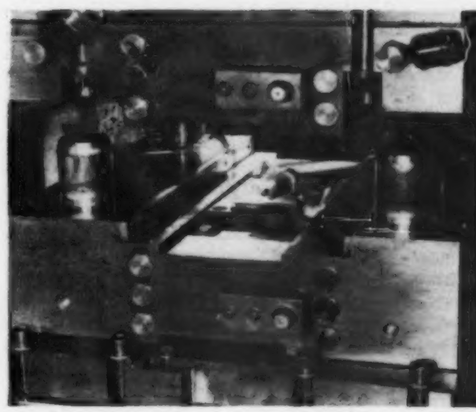
9 *Stages in manufacture of titanium compressor blade forgings*



10 *Typical dies*



11 *Blade forgings before trimming*



12 *Forging set-up*

occurred, and collapse of the die impression owing to overloading was extremely rapid.

Increasing the hardness of the forging dies produced some improvement, but it was thought that the greatest gain would be obtained if means could be found to forge the material at a higher temperature. An active search then proceeded in an effort to find a method of protecting the surface of the dummy during the heating period. A successful method would also have to provide a barrier between the titanium and the die and preferably act as a lubricant as well. Many electroplated coatings, including nickel and silver, were tried and, although both were used in a considerable number of blades, neither was fully satisfactory.

Amongst the non-metallic coatings investigated was a glass one and, as this seemed promising, a number of samples of different composition were obtained for trial purposes. These glasses were intended to fuse into a continuous protective film at a fairly low temperature, whilst having the necessary properties at the forging temperature. The method of application was to reduce the glass to a fine powder by milling and to suspend it in alcohol. The surface of the dummy was prepared by blasting or etching to provide a key, and it was then immersed in the liquid and the coating subsequently air dried. One drawback was that careful handling was necessary until the glass particles fused.

A later improvement was to incorporate the glass powder in a lacquer and apply it by spraying. When dry, the dummies could then be stored until it was convenient to forge them. Sometime afterwards, however, a proprietary brand of glass dispersion of much smaller particle size became available. This gave better coverage and adhesion and variation of this product has been used ever since. It was, of course, still necessary to use the normal forging die lubricants and for this purpose semi-colloidal graphite in white spirit or water proved suitable.

Towards the end of 1951, it became apparent that there would be advantages, from the point of view of both material supply and forging manufacture, if a change could be made to a larger size of bar. This would necessitate the manufacture of the dummies by extrusion on crank or friction screw presses and a sample of material was obtained to enable us to explore possibilities. After a number of failures, it was found that the operation was quite practicable providing some form of surface protection was employed to prevent the billet welding to the die. The normal die lubricants alone appeared to be inadequate for this purpose. As the extrusion temperature was limited to 850°, nickel plating was fairly satisfactory and has been used ever since.

Extrusion, therefore, became the standard method of dummy manufacture and arrangements were made for the bar to be supplied to us in the centreless ground condition ready for cutting to length. This arrangement also minimized trouble from surface defects caused during rolling or extrusion. Cutting the larger size of bar into slugs was troublesome at first because the abrasive wheel method employed caused damage and cracking of the cut surfaces. It was necessary to remove the damaged layers by emery dressing or turning, and this added considerably to the cost of manufacture. This trouble was finally overcome by rotation of the bar stock whilst cutting with a 0·06-in.-thick abrasive wheel, an old lathe being converted for the purpose.

Apart from ingot defects, the main trouble with early T 75 material was the fact that, owing to design considerations, only the harder batches of material were acceptable. It was found subsequently that a number of the selected batches were deficient in ductility after the various blade forging operations had been carried out, and many forgings were lost in this way. Fortunately, this trouble was cured by the introduction in 1954 of the 1½% aluminium manganese alloy which satisfied design requirements as regards tensile strength and ductility for all existing blades. Consequently, as soon as adequate supplies of the new alloy became available, the commercially pure material was abandoned. Many thousands of blades have now been made from this alloy and it is still specified wherever its properties are suitable. It has been the least troublesome of any of the titanium alloys of which we have had experience.

Unfortunately, however, the design of aircraft engines cannot remain static, and it was not long before the designers were looking for alloys of even higher strength/weight ratio. The first of these from which any quantity of blades was made was an alloy containing about 5% aluminium and 2·5% tin. This material had its own peculiarities but, since it was not used for compressor blades in production quantities, suffice it to say that these were to a large extent overcome by increasing the forging temperature to 1,040°C. and doing a large amount of work by forging to size in one operation.

A more recent development is the titanium 679 alloy which is of even higher tensile strength. It contains 11% tin, 5% zirconium, together with some aluminium and molybdenum. This material also has its own peculiarities and the best way of dealing with these is not fully established at the present time.

Finishing operations

After completing the forging and removing the excess material by hot trimming, a number of operations have to be carried out in order to bring it to the required standard of quality. These operations are mainly concerned with the blade form, the adjacent root platform and the fillet radius which connects the two. They include inspection for material and forging defects, and the removal of any such defects together with the thin layer of poor ductility which results from the various heating cycles. Forgings always carry an allowance of 0·005 in. per face to permit this. Defects are removed by dressing with emery tapes, but other methods such as wet or dry blasting, electro polishing and chemical etching are also employed.

In conclusion, it may be said that the manufacture of compressor blades in titanium is now a well-established process and that, whilst the metal, particularly some of the newer high-strength alloys, has its own peculiarities, the actual cost of manufacture compares favourably with that for the more common metals. It will be evident, however, that, owing to the high cost of the raw material and the low scrap value, the total cost of the blade is affected to a considerable degree by losses at the various stages of manufacture. Although such losses are inevitable and normally accepted in the case of low-priced materials, every effort must be made to keep them to a minimum in the case of titanium blades.

Close-to-form and close-tolerance forgings

'Modern trends in the manipulation of metals' was the theme of a conference held at Brighton last October by the Institution of Production Engineers. The present article is the second in the series on precision forging to be published in METAL TREATMENT. *It is based on the talk given by Mr. E. W. Peel, manager, Forgings Division, High Duty Alloys Ltd.*

THE FORGING INDUSTRY faced a period of change in the 1960s which will be as important as the changes in other long-established industries, brought about by the revolution in thinking which is now taking place in production engineering. During the last five years, forging engineers have realized the need to develop techniques to improve accuracy and surface finish in order to eliminate machining of forgings wherever possible. Machining techniques have advanced to such an extent that setting-up time is the major factor and the production engineer does not worry too much whether he has to remove 0·05 in. or 0·15 in. of the metal. It therefore became obvious to forging engineers that there was no virtue in getting fairly close to producing a forging which still required machining. The only way to expand the forging industry in the face of other competitive methods was to eliminate altogether the need for subsequent machining.

Over 34,000 tons of steel drop forgings were supplied during the month of June, 1960. The vast majority of these were forgings conforming to the standard dimensional tolerances of the National Association of Drop Forgers and Stampers, which control dimensional variations for thickness, die wear, offset, etc., within fairly close practical economic limits. These forgings satisfy a very substantial part of the market and are generally a first-class product.

There is a requirement, however, in an expanding field to do something more than standard die forging practice can offer and this is partly recognized in the NADFS. Notes on tolerance, where reference is made to closer-to-form forgings. The aircraft industry, for example, demands weight control on large forgings much closer than normal standard tolerances allow. Many of these requirements have been met without departing from conventional forging practice by exercising a much greater degree of control in the accuracy of die impressions, control of temperatures and 'use' shapes to reduce the thickness and offset variations in the forging operations. This has been particularly so in the case of aluminium-alloy forgings, where problems of surface deterioration due to scaling, decarburization, etc., are not encountered. Much progress has, therefore, been made in the control of aluminium-alloy forgings.

High-temperature materials

The problems encountered in making close-to-form steel forgings are more difficult to overcome; die impressions tend to lose accuracy more quickly due to the erosion and collapse of the die steel in deforming metal heated to temperatures of 900–1,100°C. If high-temperature forgings are not protected during heating and forging, scaling will result and surface condition will not be adequate to meet the needs of precision work. Much thought has been given to overcoming these difficulties.

The aim is to produce forgings in high-temperature materials, such as steel, Nimonic and titanium, comparable in every way to the dimensional and surface quality of the forgings which have been produced in light metals.

Fig. 1 shows one of the largest die forgings produced in the United Kingdom. It is a main undercarriage component supplied to Dowty for the A. V. Roe Vulcan aircraft. The overall length is 84 in., weight 932 lb. and the thickness tolerance is controlled to $+0{\cdot}1$ in. $- <0{\cdot}001$ in. This forging is machined in bores and on mating surfaces only.

In fig. 2 may be seen a fin for Armstrong Whitworth Aircraft in Hiduminium R.R.58 (DTD.731). This is not machined on the two sides, which each have three facets. The edge thickness is 0·06 in. and the centre 0·9 in. The overall length and width is 24 in. $\times$ 14½ in. respectively; thickness is held to 0·025 in.

Apart from dimensional control the problem of

1 *Main under-carriage component for Avro Vulcan aircraft*

distortion in heat treatment and subsequent slotting had to be overcome by using modified heat-treatment methods, *i.e.*, quenching in oil. By using fixtures to support the edges during the heat treatment, freedom from distortion and internal stress has been achieved. This is demonstrated by the saw cut which is part of standard release procedure for these components. No closing or opening of the saw cut is permitted.

Control fins in steel S.96 for De Havilland are illustrated in fig. 3. The tapered control surfaces are forged, allowing 0·017 in. grinding allowance on both faces. The edge thickness is 0·055 in. minimum, thickness tolerance ±0·015 in., overall length and width 10 in. × 3½ in. maximum.

Fig. 4 shows a forging in S.96 for Rolls-Royce Ltd. This forging is too large in steel to achieve the elimination of machining, but by careful control the amount of machining has been reduced to 0·093 in. on depth and 0·15 in. on plan view.

The forgings shown in figs. 1, 2 and 4 were produced using the 12,000-ton hydraulic forging press. The steel control fin (fig. 3) was produced on an 800-ton screw press. These forgings illustrate current practice using conventional methods and indicate the progress made in the reduction or elimination of machining.

Progress in forging technique

Conventional forging process There are other outstanding examples of close-to-form and close-tolerance forgings in aluminium and steel which have been made by advanced methods, but within the description of conventional die forging. In these forgings there has been a requirement to hold the die closure and die form errors within a few thousandths of an inch. These forgings are generally smaller in size than those previously mentioned.

The classic example in this field is the forged-to-size compressor blade for gas-turbine aero-engines. The need to depart from conventional machining methods arose when the first experimental compressors were released for production 20 years ago. The blade form was a curved aerofoil shape, tapered in thickness from root to tip. To machine such a shape would have involved the development of special-purpose machines resulting in delays and involving costly investment of capital. The problem was, therefore, handed to the forging engineer.

Entirely new concepts of accuracy of die sinking were involved, die impression errors being held to less than 0·001 in. Projection methods were installed for checking and a special complex technique has been developed to control accuracy of forgings throughout the manufacture.

The first precision-forged blades were in aluminium alloys but blades are now also precision forged in stainless steel, titanium alloy and Nimonic alloys.

Many millions of blade forgings have been made and the process is still supplying the vast majority of the blades now being used. Table 1 indicates the tolerances which can be held in various materials on blades from 3–6 in. long.

In order to inspect dimensionally, special-

2 *Fin in Hiduminium for A.W.A.*

purpose inspection equipment has been developed. Methods of checking in current use are: (1) liqui-column air-operated checking fixtures; (2) projection methods which scan the aerofoil profiles; and (3) conventional fixtures using dial indicators for measuring twist, bow and offset.

The checking of surface condition of blade forgings is a highly critical operation, as the surface has to be completely free from inclusions, small laps and hair-line cracks. Apart from control of shape, the surface finish for blade forgings is generally held to 17–20 micro-in. This is achieved by electrolytic polishing, followed by anodizing in the case of aluminium alloys.

To achieve the surface smoothness required in the case of stainless-steel blades, steps are taken to protect the forgings during heating to forging temperature (1,140°C.) and during forging. Various methods are used, *e.g.*, controlled furnace atmospheres, plating with various other metals, the use of special lubricants, *e.g.*, colloidal glass.

The resultant product passes the very stringent inspection processes which the aero-engine industry

4 *Forging in S.96 for Rolls-Royce*

TABLE 1 *Blade tolerances*

Blade length (inches)	Thickness (inches)		Disposition (in./1-in. length)	Angular displacement (degrees)	Profile tolerance (inches)
3	Hiduminium Steel and titanium	+0·005 +0·007	0·002	½	+0·0015 to give smooth surface
7	Hiduminium Steel and titanium	+0·008 +0·010	0·002	½	+0·002
12	Hiduminium Steel and titanium	+0·012 +0·015	0·002	½	+0·003
16	Hiduminium Steel and titanium	+0·018 +0·025	0·002	½	+0·004

The foregoing is given as a general guide only and different tolerances may be applied to blades with special characteristics.

3 *Control fin in steel S.96 for De Havilland*

demands. Methods developed under these conditions have now been applied to other forgings and eliminate machining.

Limitations of conventional forgings Although much progress has been made, there are limitations in shape, tolerance, etc., which cannot be avoided by the conventional method of forging between two dies. These limiting factors can be summarized as follows:

1. The need to fill up undercuts;

2. Draft angle must be allowed to enable removal from dies;

3. Fillet radii cannot be reduced beyond certain limits, compatible with the internal shape of the forging;

4. Die design must avoid weak projections and features which result in die failure;
5. Die-closure tolerances must be allowed, and variations can be expected within these tolerances;
6. Die wear results from the movement of metal over the die faces into the scrap gutter.

New approaches to forging problems (light alloys)

Split die forgings In order to overcome these limitations, much thought has been given to tool design and to developing different forging machines such as hydraulic or mechanical presses having side-acting rams. The ' hot brass ' pressing industry has used side-acting tooling for many years to make forgings such as water taps and valves. This approach has been developed on a much larger scale in the United States by the Cameron Iron Company, who have installed an 11,000-ton hydraulic press with two 6,000-ton side-acting rams. They use these for the manufacture of large hollow forgings in steel for the oil well and similar industries.

This approach is now being applied to many forgings in aluminium and magnesium alloys with great success.

The advantages gained are: (1) die-closure tolerances are dependent only on how tightly the dies are held together during forging; (2) undercuts can be produced; (3) die wear can be reduced because of minimum metal flow; and (4) flashline effects (grain) are greatly reduced.

Fig. 5 shows typical samples of split die forgings and fig. 6 shows a pump body forging being removed from the tools.

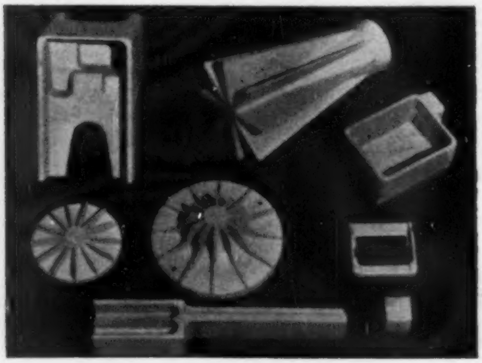
Enclosed die forgings This method also has been developed for aluminium and magnesium alloys. In this process the metal is totally enclosed in the dies, the punch member sealing off the cavity. On the descent of the punch the metal is forced to flow into the various cavities in the top and bottom members of the die. The only escape for the metal is provided in the form of vertical vents of small cross-section, thus the control of volumes is extremely important. Forgings without draft and flash-line effects can be produced by this method with small fillet radii.

Errors of shape are limited only by the accuracy of the impressions of the top and bottom members.

6 *Pump body forging being removed from the tools*

5 *Typical examples of split die forgings*

7 *Enclosed die forgings in aluminium alloys*



8 *Sole-plate in titanium alloy*

Die-closure tolerances are, however, still required but can be held to approximately +0·02 in.

In the case of the single-sided impeller shown in fig. 7, the important form is on one side only so that the base is usually thickened up and the requirement is to achieve fully-made-up fins. The base is reduced to final thickness by a simple turning operation. Dimensions achieved on this impeller are: Vanes: thickness ±0·003; disposition 0·005. Variation on faces between vanes: 0·004.

The impeller type of component is especially suited to this type of technique as the various dies and punches can be finished by turning and grinding. Other plan shapes can be accommodated but should be kept to relatively simple geometric forms such as rectangles with minor variations.

The limit in plan area is governed by the size of the press available and the web thickness required. The largest practical area at the moment is 150 sq. in. with a web thickness in this case of 0·375 in., needing a pressure of 3,000 tons.

One of the features of this process is the die life that has been achieved. In the case of the buckle component illustrated, the die life has reached 200,000 forgings.

Cold forging (*light alloys*) This technique is receiving much attention at the present time due to the accuracy which can be achieved (0·0005 in.) and the resultant possible savings in material and in machining hours. In the case of aluminium alloys, the cold impact extrusion technique is the standard method of making collapsible tooth-paste tubes and similar articles, many hundreds of thousands weekly. The process is being applied to thicker-walled components, and there are also various developments of the process in the steel industry which will be touched upon later.

Internal and external keys and keyways can easily be provided with stepped bores and formation on the base of the extrusion. Tolerances on internal and external diameters of 0·0005 in. can be achieved. The surface condition is excellent, being brightly polished.

In the process, a cold slug of metal is placed in the container and the punch is forced into this slug causing the metal to extrude back over the punch, and sometimes forward into a die cavity. In impact extrusion the punch travels at a high speed, and the extrusion operation is completed in a fraction of a second. Cold-forging operations can, however, be carried out at slower speeds, for such operations as upsetting and heading. Stresses in tool members are high at the point of impact and tool strength and finish are therefore critical. For cold-impact extrusion, special-purpose presses have been developed which have very robust frame construction and sturdy punch guides.

The process is suitable for production runs of 10,000 upwards, bearing in mind the cost of tool manufacture and development, and setting up costs.

Aluminium alloys are well suited to the cold-forging process and an increasing range of components is being produced economically by this means.

Progress in close-to-form forgings in steel and other materials Much attention is being given to the development of both hot and cold methods of forging which will produce clean, precise steel forgings. Compressor-blade forgings are currently being produced close-to-form in stainless steel, Nimonic and titanium alloys to a degree of accuracy and surface finish which compares favourably with that produced in aluminium alloys.

Fig. 8 shows a sole-plate in titanium alloy. The edge thickness is held to 0·035 in. +0·015 in.-<0·001 in. Due to the curved form this would be a difficult machining operation and close-to-form forging has proved to be a very convenient method showing as it does a considerable material saving.

Cold forging (*steel*) Cold forging in steel is becoming increasingly important and many components for the automobile industry formerly finished by machining are now being made more by, or planned for, this process. The extrusion technique is being applied to hollow components, but in addition such items as small gears and splined components are being successfully produced. Several suppliers of forgings to the automobile industry have already set up production lines of special plant for cold forging, and the automobile manufacturers themselves are introducing cold-forging presses into their production lines.

The plant required for these operations is usually specially designed or adapted, but it is possible to use conventional presses for some operations where speed of deformation is not important. The automobile manufacturer seldom requires the highest speed equipment, as this is usually in excess of the hourly requirement. One leading company is purchasing simple presses specially designed to make one component per press, at the speeds the production line requires.

The supplier to the industry, however, is interested in the high production rates which the advanced press equipment now available can offer—at a correspondingly higher capital investment.

Tool design and know-how is very important in the successful development of this process, as the stresses set up at the point of contact are greater. Tools are generally made in 12% chromium, 2% carbon hardened to 57–58 Rockwell C and finished to a smoothness of 4 micro-in.

Preparation and lubrication of slugs is important and various special treatments such as phosphating followed by impregnation with various lubricants have been developed. Great savings in materials are possible as waste in machining chips is largely avoided.

Future trends

In looking into the future, trends of development are already apparent. On the one hand more advanced methods will be used to make close-to-form forgings in the cheaper carbon steels, *e.g.*, cold forging, with resultant savings in material, time, machines and factory space.

On the other hand, newer and more difficult materials are being specified which are becoming increasingly difficult to manipulate by conventional methods. Materials such as the latest Nimonic alloys, beryllium, zirconium, molybdenum and many other new materials will be handled in the next decade.

New methods are being developed in order to deform these metals, which show great promise in the precision and accuracy which can be attained. Very high speed, high energy rate, methods are being developed by Convair in the U.S.A., using a new machine called the Dynapak. This machine can deliver accurately controlled energy up to 450,000 ft./lb. at velocities up to 2,000 in./sec.

Another process under development uses an explosive charge to provide the energy for deformation.

With the advanced methods becoming available, profound changes in manufacturing technology may be expected in the near future, offering the designer new freedom in shapes, and the production engineer the accuracy he has long required.

Reflectors save heat during forging

THE PRIMARY CAUSE of heat loss from forgings at temperatures above 800°C. is radiation, and it should therefore be possible to reduce this loss by reflecting the heat back on to the forging.

Experiments in BISRA's Sheffield laboratories have shown that by surrounding a hot bar with a polished aluminium tube the rate of cooling, during forging, could be reduced by a factor of one and a half. This means that a forging which would normally require three heats could be completed in two—a considerable saving.

In the research the reflecting material was formed into a tube completely surrounding the bar to be tested. Three materials—aluminium sheet, stainless steel and mild steel sheet—were tried; aluminium was proved to have better reflectivity, and to retain its reflectivity longer and at higher temperatures than the stainless steel; if it were raised to a temperature above 300°C., the stainless steel oxidized quickly, became dull and lost its efficiency as a reflector, whereas aluminium could be raised to a temperature of 600°C. without noticeable deterioration. The reflectivity of mild steel was inferior to the two other materials.

The diameter of the reflecting tube determines the temperature of the reflector, and it also has an important effect on the conservation of heat in the forging: the smaller the diameter, the slower is the loss of heat. For this reason the ratio of tube diameter to bar diameter was tabulated in several experiments, and it was found that for aluminium sheet a tube about twice the diameter of the forged bar gave optimum heat conservation consistent with maintaining the reflectivity of the tube.

Forging experiments, with and without reflectors, were made on the BISRA experimental forge, and the results compared. The ingots used were of 8-in. square section. On the free side of the press the shield was an aluminium tube of 15 in. diameter; on the manipulator side a tube was made up of segments with a radius of 9½ in., so that segments could be removed as the manipulator was moved into the press after each forging pass. The overall diameter of 19 in. on this side allowed the manipulator gripping jaws to be entered inside the shield.

Packaged boilers installed in new Wiggins' factory

Largest single-packaged boiler installation in this country has just been completed by G.W.B. Furnaces Ltd., Dudley, Worcs. It is part of a multi-million-pound project being carried out by Henry Wiggin & Co. Ltd. at their Hereford Works where five Model 600 Powermaster packaged boilers have recently been put into service. The new factory, which covers an area of some 52 acres, will absorb Wiggin companies at present based in Scotland and Birmingham. The new boilers will mainly be used for space heating, though a little steam will be raised for process work.

Effect of carbide stringers on the distortion of die steels during heat treatment

K. SACHS, Ph.D., M.Sc., A.I.M.

The causes and mechanism of distortion of die steels during heat treatment, the influence of the structure of the steel and in particular the part played by carbide stringers, are studied. The author is Head of Research Metallurgy Section, G.K.N. Group Research Laboratory, Wolverhampton, and his article will be continued in future issues

continued from last month

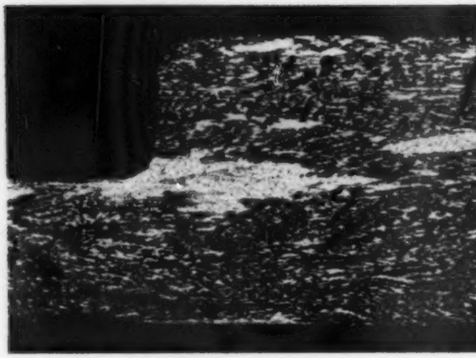
The distortion of the specimen is the resultant of the interaction of a number of factors: temperature gradients during quenching, variations in the amount of retained austenite in different zones, and differences in the anisotropic dilatations of adjacent zones. It was not possible to determine the influence of each factor separately, but the two main causes of distortion, the geometry of the specimens and differences in carbide structure, could be distinguished by a careful analysis of the distortion behaviour. In most cases specimen *B* had a uniform structure throughout and was expected to distort only slightly, under the influence of geometrical factors. Specimens *A* and *C*, on the other hand, consisted of zones of different structure and differential dilatation behaviour was expected to play a much greater part in their distortion; moreover, they were disposed in such a manner, relative to the different zones in the forged bar, that they distorted in opposite directions. In some billets specimen *B* showed appreciable distortion but that was generally associated with penetration of the finer carbide structure into the central zone. Another cause of severe distortion in specimen *B* is the presence of very coarse segregate particles in the internal corner where the change in section occurs. This is illustrated in fig. 29; the carbide structure in the dendritic and segregated zones are compared in figs. 30 and 31. Such a sharp change in structure at the point where it has most effect on the bending of the specimen naturally finds its expression in an abnormally high distortion index.

Vertical and horizontal quenching[19, 20]

It is generally recommended that long pieces of steel should be quenched vertically to reduce distortion. It is often convenient, however, to place heavy dies flat on a perforated plate which is then lowered into the oil bath by means of a chain and

Table 6 *Distortion figures after oil quenching from 960°C.*

Furnace	Quench	Distortion					
		(Fig. 26) Specimen					(Fig. 27) Specimen *D*
		1	2	3	4	5	
Salt bath	Vertical	+6·1	+4·6	+5·9	—6·4	+ 4·0	+1·4
Tube furnace	,,	+2·9	+9·2	+3·4	—4·5	+10·0	
,,	,,	+5·5	+6·2	+7·3	—5·1	+ 7·7	
,,	,,	+4·9	+5·7	+6·8	—6·9	+ 2·8	
		(Mean) +4·4	+7·0	+5·8	—5·5	+ 6·8	+6·4
,,	Horizontal	+3·0	+5·5	+4·8	—2·1	+ 2·8	
,,	,,	+2·3	+3·0	+5·7	—6·9	+ 5·8	
		(Mean) +2·6	+4·7	+5·3	—4·5	+ 4·3	+8·1



29 *Coarse segregate at internal corner of distortion, specimen B* ×4

pulley block. The response of ' simulated dies ' to vertical and horizontal quenching was tested on the angle and T-sections and on one specimen (type *D*) of the alternative design. The specimens were oil quenched from 960°C. and the distortion figures are summarized in Table 6.

The effect of quenching direction is negligible. The greatest difference is between hardening from a salt bath and a tube furnace, which points to the effect of even small longitudinal temperature variations, an effect to which type *D* is, of course, more susceptible than the old type of specimen with no change of section along the length. The latter specimens show a little less distortion in horizontal than in vertical quenching, but the difference is not significant in view of the large scatter. The ' new ' specimen has distorted a little more in horizontal than in vertical quenching but the effect is marginal compared with that of the hardening furnace, and this, in turn, is much smaller than others to be discussed in later sections.

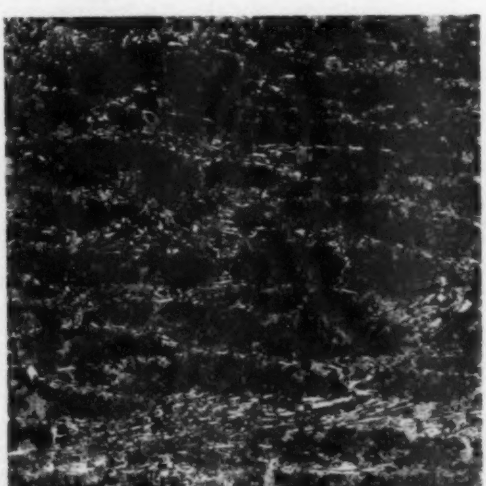
It can be concluded that for comparatively light sections like those employed in these experiments, it makes little difference how they are immersed in the quenching tank. Heavier dies and components may, however, be much more sensitive to quenching methods.

Non-uniform hardening temperature[14, 20]

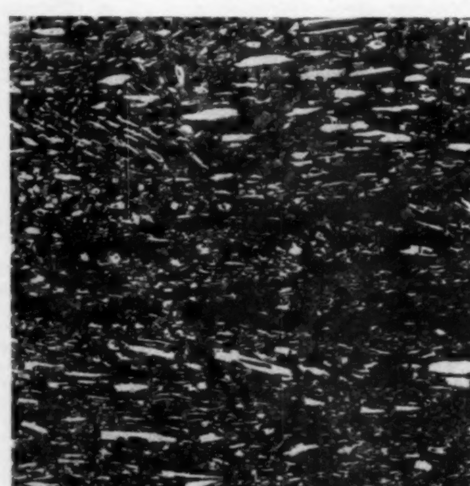
Distortion specimens of both designs were quenched from uniform and linearly varying hardening temperatures. In order to impose a linear temperature gradient a special furnace was constructed. It consisted of a mullite tube, 2½ in. dia. and 2 ft. 6 in. long, with a special winding and circuit intended to give a uniform temperature over 8 in. at least, as well as the possibility of imposing an accurately controlled temperature gradient over this length. The tube was gradient wound over the central 14 in. with 16 s.w.g. Brightray wire, varying from 10 turns in. at the ends to 5 turns in. in the middle.

The circuit, illustrated in fig. 32, is a modification of Weaver's[19] creep test furnace. The mains current passes through a potentiometric temperature controller which records the readings of two thermocouples, in the centre and at one end of the 8-in. heating zone; one of these thermocouples actuates the control relay which switches the current on and

30 *Carbide structure in dendritic zone of* **29** ×100



31 *Carbide structure in segregated zone of* **29** ×100



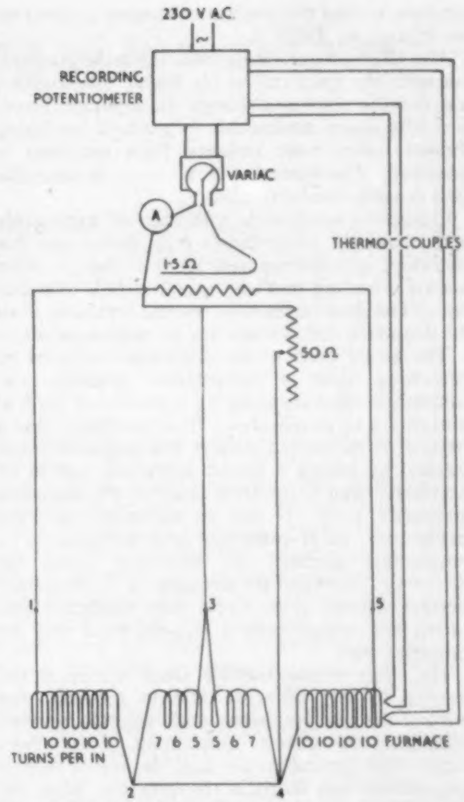
off to keep the furnace temperature within $\pm$ 7°C. of the setting. The current is reduced to a maximum of about 10 amps by a Variac transformer, whose output is connected to the furnace winding, which is split into three sections. The two outer sections are connected in parallel, the current being supplied from the Variac through an adjustable rheostat used as a potentiometer, so that movement of the sliding contact adjusts the proportions of the total current passing through the two outer sections. The central section, which is wound in two parts, is in series with the two outer sections and the current is controlled by a shunt resistance. This arrangement of the windings and tappings has the advantage that there is no potential difference at any of the joints in the split winding; thus there is no danger of sparking and no need for large gaps at these joints.

The thermocouples are threaded in four-hole insulator beads, with the hot junctions 4 in. apart; they can slide in a mullite sheath so that the temperature at any point along the 8-in. heat-treatment zone can be recorded. It was found that the temperature variation over the 8-in. zone could be held down to 3°C.; better results might be anticipated from a furnace with a longer winding. The same accuracy was maintained in treatments with temperature gradients of 20°C. and 50°C. applied over the 8-in. specimen; steeper gradients were more difficult to control and could not be consistently reproduced in treatments on different specimens. The rather wide fluctuation with time was not thought likely to affect the distortion in heat treatment.

Although the tube diameter was just large enough to take all the specimens at once, the heat capacity of the furnace was so low that the removal of a specimen for quenching affected the temperature considerably and it took some time to settle down again. The specimens were therefore hardened separately. Some hardening treatments were carried out in a neutral salt bath which was also used for the martempering experiments described later.

Hardening treatments were carried out with uniform temperature distribution, and with gradients of 20°C. and 50°C. over the 8-in. length of the specimens. The effect of reversing the gradient was studied on the type of specimen shown in fig. 27; the thick and thin ends of the specimen were at the higher temperature in different quenching treatments. The results are presented in Table 7.

The influence of the temperature gradient on the distortion of the angle- and T-sections was less than the random fluctuation in distortion figures. The design of the specimen was modified to make it more sensitive to longitudinal temperature gradients



32 *Circuit diagram of gradient furnace*

and the new type of specimen responded quite sharply to such hardening conditions.

The results for this type of specimen quenched from the salt bath, where the temperature distribution is as uniform as one can reasonably hope for, is characteristic of the distortion behaviour in this test. Specimen *B*, which distorts predominantly under the influence of geometrical stresses, shows very little distortion, while specimens *A* and *C* distorted 0·014 in. and 0·012 in. respectively in opposite directions. When such specimens are quenched from a temperature gradient falling by 50°C. from the thin end to the thick end, specimen *B* distorts appreciably, the distortion of specimen *C* is practically doubled, while the negative distortion of specimen *A* is virtually cancelled. The simplest way of expressing the effect of the gradient hardening is in terms of a positive distortion component of 0·012 in. to 0·015 in. added to the distortion figures associated with quenching from a uniform tem-

perature, so that the resultant distortion is given by the figures in Table 7.

The effect of quenching from the same gradient, but with the thick end at the higher temperature, was broadly similar, although the separate distortion component attributable to gradient hardening showed rather more variation from specimen to specimen. The treatment in the works furnace also gave roughly similar results.

Quenching a specimen with one end appreciably hotter than the other causes it to distort and this distortion is superimposed on that due to other causes, including that due to the carbide distribution. The final dimensions are the resultant of all the distortion components due to various causes.

The actual cause of the distortion produced by quenching from a temperature gradient was established experimentally by a process of gradual elimination of possibilities. The possibility that a gradient of quenching stresses was responsible was checked by testing a ferritic specimen, cast in an electrolytic iron to the same shape as the distortion specimens in fig. 27 and an austenitic specimen cast in 18/8. After quenching from the usual 50°C. temperature gradient the distortion index for ferrite was +1·0 and for austenite −1·7. Evidently thermal stresses alone cause only negligible distortion and transformation stresses must play an important part.

The effect of pure transformation stresses in the absence of any influence due to carbides was checked on cast specimens with analyses roughly resembling the matrix composition of the high-carbon high-chromium die steel; in one of them a little carbide was found in the structure, while the other contained no primary carbide. The distortion of these specimens was roughly similar to that found in specimens cut from forged bar and quenched from a temperature gradient.

The amount of retained austenite in various positions on gradient-quenched specimens was estimated by means of X-rays on specimens cut from forged bars of the die steel and on specimens cast in the 'matrix' composition. The carbide-free specimen had no retained austenite, the one with a little carbide had about 7% at the hot end only while the die steel specimens varied from 11–16% at the cold end to 18–29% at the hot end.

Since distortion due to quenching from a gradient is independent of the presence of carbides or retained austenite, but only occurs in material that passes through the γ/α transformation, we can reasonably attribute it to transformation stresses. The cold end of the bar reaches the M_s temperature first and the martensite transformation travels along the length of the bar. Complex stresses are set up in the area where the section changes sharply and distort the specimen severely.

The practical conclusion is that the temperature distribution in the hardening furnace should be uniform. This may appear a little obvious—it hardly seems worth while to carry out elaborate distortion experiments in special furnaces if the outcome is merely the reiteration of the importance of carrying out delicate heat-treatment operations in furnaces that are suited to the job. Nevertheless, a decision to modify or replace an existing furnace involves a fair amount of money; this can be faced more cheerfully if there is empirical evidence to confirm the benefits to be derived from the change in addition to theoretical prediction, however confident.

to be continued

Group visit to Moscow

Forty-eight directors of firms in the Gauge and Tool Makers' Association will be taking part in a group visit to Moscow which the Association is organizing during the forthcoming British Trade Fair in that city (May 19 to June 4, 1961).

TABLE 7 *The effect of temperature gradient in hardening*

Furnace	Temperature gradient °C.	Distortion										
		'Old' type specimen (fig. 26)					Alternative specimen (fig. 27)					
							Bar AAK			Bar ADX (B)		
		1	2	3	4	5	*A*	*B*	*C*	*A*	*B*	*C*
Tube* ..	0	+4·4	+ 7·0	+ 5·8	—5·5	+6·8						
Salt bath ..	0	+6·1	+ 4·6	+ 5·9	—6·4	+4·0	—13·9	+ 1·8	+12·5	—14·0	+3·7	+12·3
Tube	20	+4·1	+10·5	+11·2	—6·2	+6·0						
Tube	50	+5·9	+ 7·2	+ 5·8	—6·6	+4·7						
	Thin end 985, thick end 935						— 2·0	+17·1	+24·7			
	Thick end 985, thin end 935						— 2·5	+ 2·8	+18·7	— 3·5	+8·0	+14·4
Large gas-fired works furnace	About 50	+4·2	+ 1·9	+ 2·3	—7·5	+5·1	— 6·6	+ 5·5	+22·5			

*Mean values from Table 6

Precipitation hardening of low-carbon steel

MIRKO KLESNIL and PŘEMYSL RYŠ

Structural changes of a super-saturated solution of alpha iron in heat-treating experiments with a low-carbon steel were examined metallographically by optical and electron microscopy. The work was first reported in ' Hutnické Listy,' 1960 (11). *The authors are with the Laboratory for the Study of the Properties of Metals, at the Czechoslovak Academy of Sciences in Brno*

THE BREAKDOWN of a super-saturated solid solution through precipitation is a complicated phase transformation of considerable technical importance. Over a number of years it has been used in particular for the precipitation hardening of non-polymorphous alloys of non-ferrous metals such as light alloys and certain bronzes, but in recent years it has been applied to a significant degree to heat-resisting materials such as austenitic steels and nickel alloys.

According to the modern theory of hardening, evolved for substitution solid solutions, a distintegrating, super-saturated, solid solution passes through certain characteristic states in accordance with certain laws. At sufficiently low temperatures in relation to the temperature of the appropriate point on the solubility change curve the homogeneous, solid solution is transformed into an inhomogeneous one, with the formation of Guinier-Preston zones enriched in the element with which the solid solution is saturated; these zones are coherent with the lattice of the solid solution of the matrix. After a longer period, or on raising the temperature, these zones are further transformed into precipitates of a transition phase, which are finally changed into an equilibrium phase. Subsequently, in essence only those structural changes take place, during which coalescence and coagulation of the initially fine precipitates take place. During the course of the precipitation there are marked changes in the mechanical and physical properties of the alloy. By nature the transformation is one of diffusion, and takes place through the formation of nuclei of a new phase and their subsequent growth.

In alloys of iron, carbon and nitrogen are capable of forming a solid solution in alpha iron; the solubility of both these elements decreases with falling temperature, so that favourable conditions are created for the formation of a super-saturated solid solution and its breakdown through precipitation. And although in this instance it is a matter of an interstitial solid solution, the basic system of the sequence of precipitation is the same as in substitutional solid solutions. The homogeneous solid solution transforms into an inhomogeneous one with an increased concentration of interstitial atoms in areas with lattice faults, such as dislocations and vacancies and, then later on, dependent on the level of the temperature of the process, the actual precipitation of the transitional or equilibrium phases takes place. The interstitial solid solution of C and N in alpha iron, and its breakdown through precipitation, are the cause of certain characteristic properties of carbon steels, such as the existence of a marked yield strength, blue brittleness and various sorts of ageing, for instance. The study of precipitation processes in mild steels has importance not only for a more complete knowledge of the properties mentioned, but also for a deeper knowledge of the process itself.

Precipitation of carbides from ferrite

During the gradual cooling of low-carbon steel from the A_1 temperature, carbon is precipitated through the effect of its decreasing solubility in the ferrite as iron carbide unless, of course, cooling proceeds so slowly that the equilibrium state can be reached. The carbide is precipitated preferentially on the grain boundaries, and is called tertiary cementite.

By rapid cooling from the eutectoid temperature it is possible to obtain a metastable, super-saturated solid solution of carbon in alpha iron with a content of 0·018% C. Since even at normal temperatures the diffusion rate of carbon in alpha iron is sufficiently high, the carbon diffuses into the lattice defects, so that within a relatively short time an inhomogeneous solid solution is created; at elevated temperatures precipitation of the carbon

in the form of microscopic carbides gradually takes place.

Precipitation of carbon from the solid solution of alpha iron is accompanied by changes in the physical and mechanical properties. There are very marked changes in the damping capacity and electrical conductivity; therefore measurement of these values serves as a good method of accurate quantitative determination of the carbon content of the alpha solid solution.[1] The mechanical properties, such as hardness and yield strength, have differing sequences at given temperatures in relation to the time of precipitation, *i.e.* to the carbon content of the ferrite; as distinct from the time sequence of the physical property values, after a certain time they attain maximum values, dependent on the precipitation temperature, as Davenport and Bain have shown.[2]

The type of precipitate which forms from the breakdown of the super-saturated, alpha solid solution in carbon steels is dependent on the precipitation temperature. Within the range of precipitation temperatures from 100° to almost 300°C., by means of electron diffractography the presence of the hexagonal ε-carbide has been shown,[3, 4, 5] which also precipitates in the first stage of martempering.[6] In both instances the metastable ε-carbide probably precipitates, because its lattice is more closely related to the lattice of the ferrite than the lattice of orthorhombic cementite. The spacing of the [101] planes of ε-carbide is very close to the spacing of the [101] planes of ferrite, and the arrangement of the atoms in these planes is analogous. At higher temperatures the ε-carbide changes into the stable, orthorhombic carbide, cementite, the presence of which may be reliably indicated after heating at a temperature of 300°C.

The changes in the structure can also be followed metallographically. On account of its low power of resolution the optical microscope cannot provide closer information, but electron microscopy makes it possible to establish the morphological changes even in the initial stages of precipitation. According to the theoretical treatises of Wert[7] and Zener,[8] carbon is precipitated in the form of the globular carbide Fe_3C; Radawich and Wert[9] tried to confirm this opinion. The much more detailed work of Tsou, Nutting and Menter[3] did not confirm this opinion. On heating a super-saturated solution of alpha iron to a temperature of 100°C., these authors found lamellar precipitates of ε-carbide of about 500 Å in thickness. Within the temperature range from 200–300°C. the ε-carbide is gradually transformed into the iron carbide Fe_3C, which likewise precipitates in the form of lamellae; the characteristic lamellar shape is preserved even after heating to a temperature of up to 500°C. Within broad outlines other authors also came to similar conclusions.[4, 10, 11, 12]

For knowledge of the mechanism of precipitation, of great importance are the measurements of Wert[7] and Gruhl[13] on specimens which were cooled from a temperature of 700°C. and then preliminarily hardened at a temperature of 20°C. Wert investigated the changes in damping capacity after heating to a temperature of 50°C., and Gruhl the change in the electrical conductivity after heating to a temperature of 100–350°C. in specimens hardened at a temperature of 20°C. The results of measurement showed that the precipitation of carbon takes place much more rapidly than without preceding hardening.

The effects of reconstitution of the solid solution during short-term annealing after preceding hardening at considerably lower temperatures were described by Gruhl,[13] and Köster, Geller and Kuntze.[14] If a specimen which has been hardened at a temperature of 20°C. is annealed for a short time at a temperature of 100–200°C., to a certain extent a homogeneous solid solution is again formed, which is manifested by an increase in electrical resistivity and a fall in hardness. On the basis of the treatises of Becker[15] it may be suggested that the nuclei of the carbides have dimensions corresponding to the critical size of nucleus for the hardening temperature employed. Nuclei which have formed at lower temperatures can again dissolve during short-term annealing at a higher temperature, since their dimension is less than the critical size of the nuclei corresponding to this temperature.[1, 2, 13]

The precipitation processes already described, which take place in a solid solution of alpha iron super-saturated with carbon, may be substantially accelerated by mechanical working. The physical and mechanical properties change in the same way as in the preceding instance, but the changes take place much more rapidly[13] and the mechanical properties attain higher values.[2] At the same time a stabilizing influence of the preceding deformation is manifested, consisting in the fact that the mechanical properties do not change noticeably under the influence of further heating. It is possible to assume that the specific properties which are shown by the super-saturated solid solution of carbon in alpha iron during hardening after preceding deformation are dependent on the high number of dislocations which are formed by the mechanical working. The carbon atoms contained in the lattice of alpha iron can therefore reach the dislocations along short diffusion paths, so that the number of nuclei is considerably greater, and their average size is smaller. Hardening understandably proceeds more rapidly, and attains much

higher mechanical values at a selected hardening temperature.

Material and method of the experiments

For the investigation of the changes in the structure of a super-saturated solution of alpha iron, a low-carbon steel was chosen with a content of 0·05% C. and 0·0042% N. First of all the steel was heated at a temperature of 100°C. for a period of 1 h. with subsequent gradual cooling in the furnace over a period of 24 h. This treatment served to produce a suitable grain size of about 0·01 mm. The specimens had the form of platelets of 4 × 15 × 40 mm. The specimens prepared in this way were heated at a temperature of 700°C. for a period of 1 h. and rapidly quenched in water at a temperature of 20°C. The specimens were electrolytically polished in an electrolyte of composition: 225 ml. CH_3COOH, 5 ml. H_2O and 20 ml. $HClO_4$ and etched for a period of 30 sec. in 2% nital.

For hardening, temperatures were chosen within the range from 23–128°C. With the use of a thermostat the specimens were heated up to the selected temperatures which were maintained within the limits of ±0·05°C. The hardness was measured by the Vickers method with a 10-kp. load. The resultant hardness values were determined as the arithmetic mean of ten measurements. For this arithmetic mean in all instances the mean square error was calculated, which varied within the limits of ± (0·5–2) hardness units.

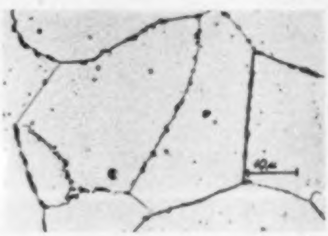
The structural changes were investigated matallographically in a Zeiss-Neophot optical microscope and in a Tesla BS-242, table-type, electron microscope, using two-stage, colodion-carbon replicas shadowed with gold and palladium.

Experimental results

The structure of a specimen of the steel cooled in the furnace is formed of ferritic grains and tertiary cementite, predominantly precipitated on the grain boundaries (fig. 1). After rapid quenching in water from a temperature of 700°C. only grains of ferrite remain in the structure with very smooth and clean grain boundaries (fig. 2). After 13 hours' heating at 50°C. it is possible to observe a difference in the etchability of the individual grains (fig. 3), but concerning the character of the changes in their structure the micrograph from the optical microscope cannot disclose the required information.

The sequence of precipitation at temperatures of 23, 50, 77, 97 and 128°C. in specimens quenched in water was investigated by measurements of the hardness at normal temperature, and metallographically by means of the electron microscope, in relation to time. The time sequence of the hardness after hardening at the temperatures mentioned is shown in fig. 4.

All the hardness curves have a form which is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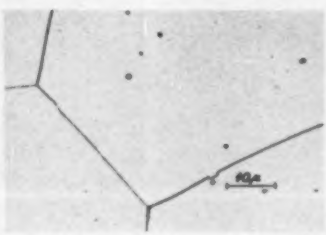
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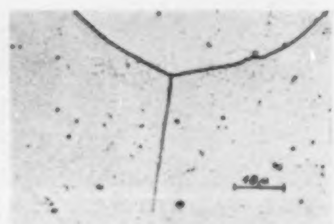
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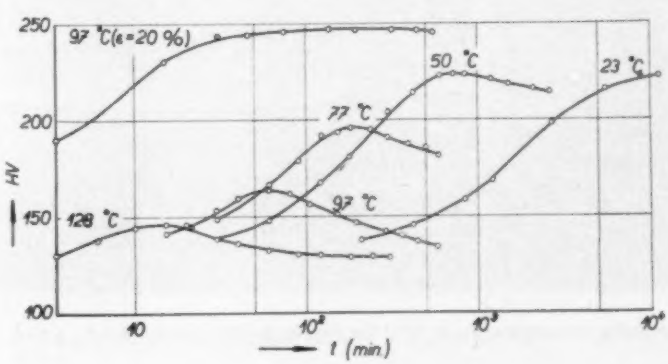
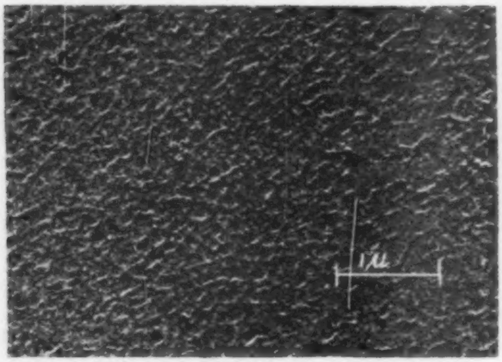
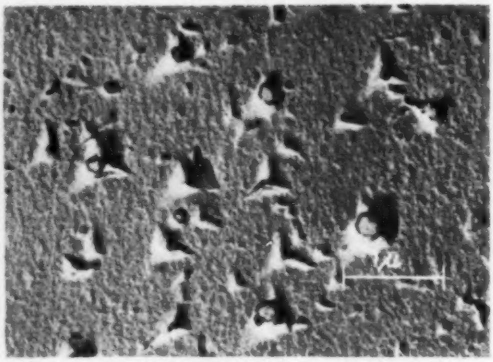
ABOVE **1** *Tertiary cementite precipitated on the boundaries of the ferritic grains;* **2** *Boundaries of the ferritic grains after rapid quenching of a specimen in water from a temperature of 700°C.;* **3** *Appearance of the ferritic grains after hardening at a temperature of 50°C. for a period of 16 h.*



LEFT **4** *Time plot of hardness for hardening at various temperatures*

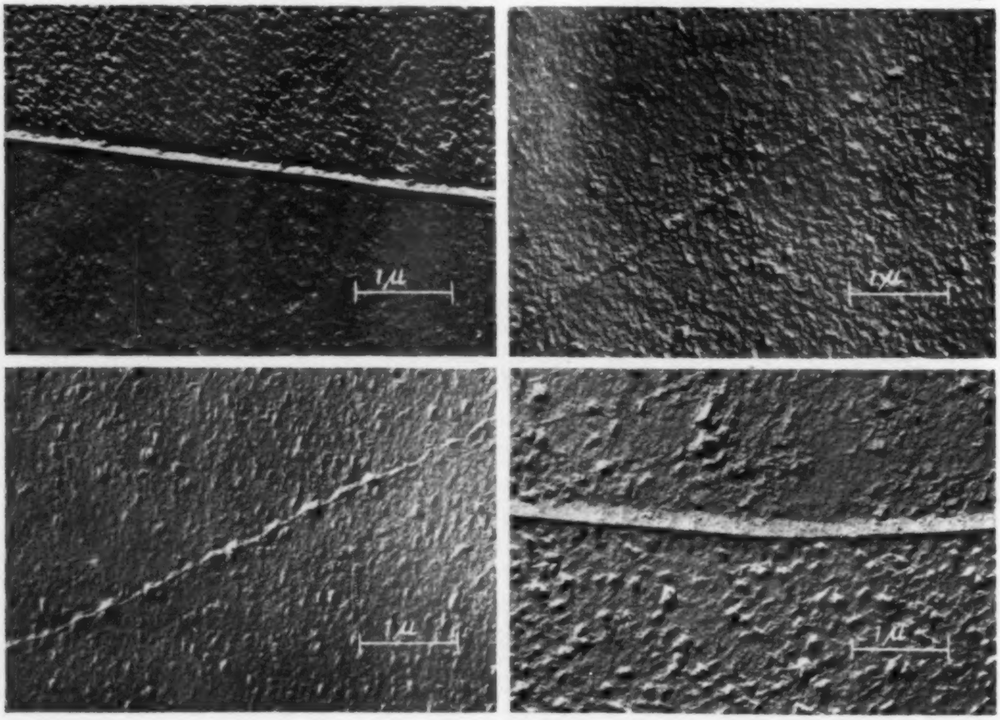


5 *Structure of ferrite after hardening at a temperature of 23°C. for a period of 16 h.*



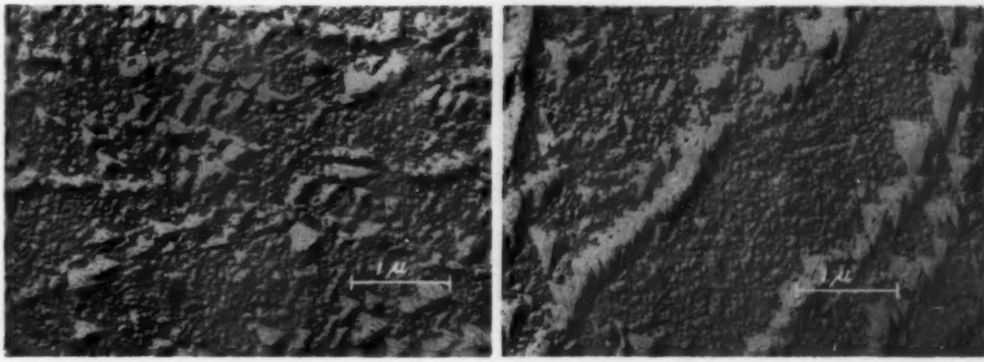
9 *Structure of a ferritic grain after hardening at a temperature of 250°C. for a period of 2 h.*

a b



c d

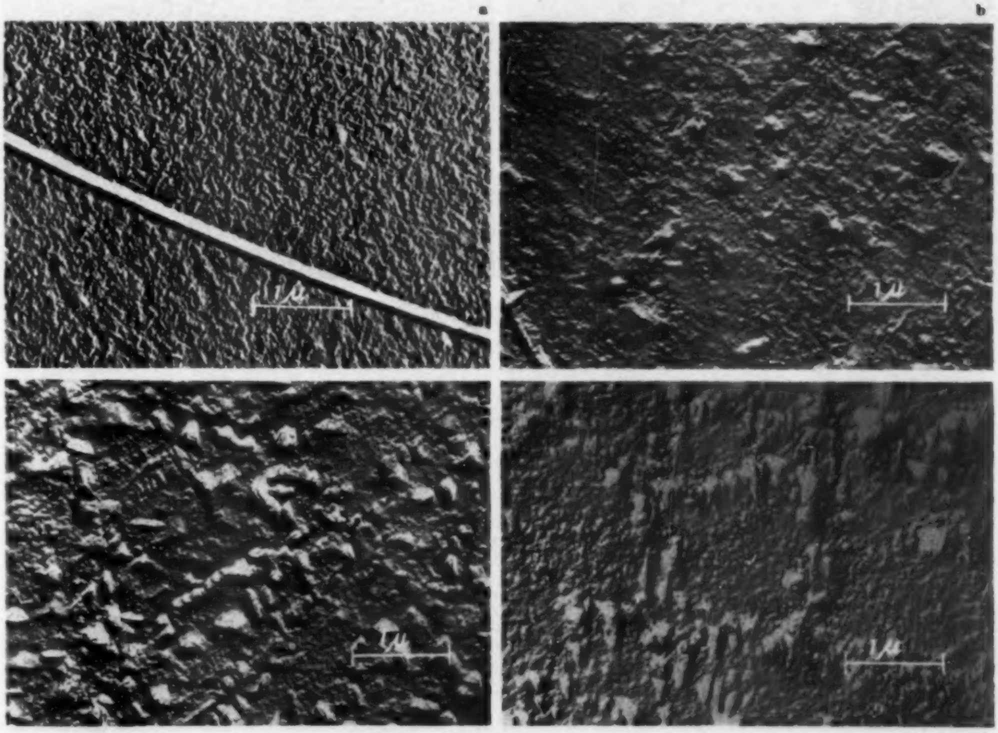
6 *Structure of ferritic grains during hardening at a temperature of 50°C. for various periods:* **a** 1 *h.;* **b** 6 *h.;* **c** 13 *h.;* **d** 41 *h.*



a b

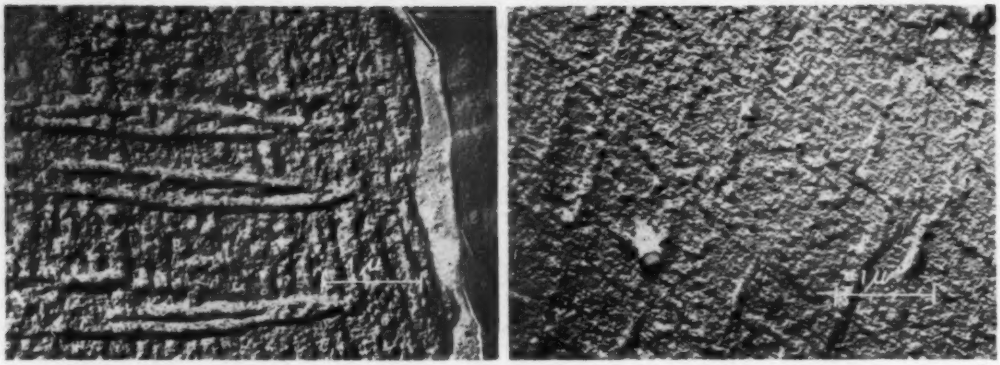
8a and b *Structure of ferrite after hardening at a temperature of* 128°C. *for a period of* 9 *h.*

a b



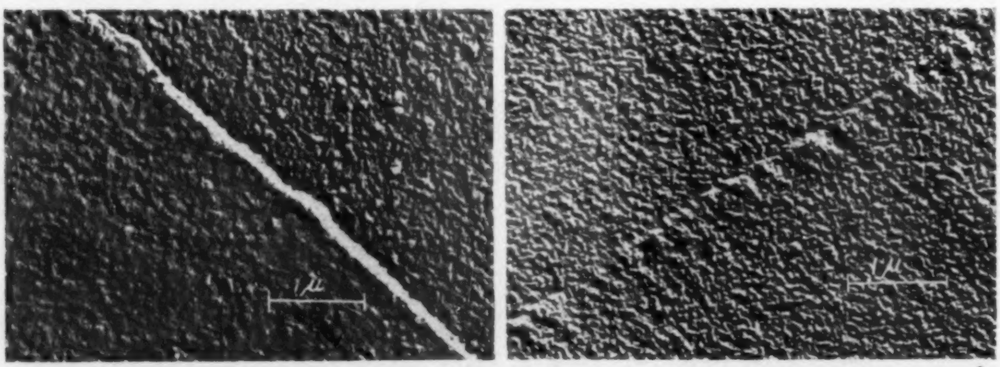
c d

7 *Structure of ferritic grains after hardening at a temperature of* 97°C. *for various periods:* **a** 30 *min.;* **b** 1 *h.;* **c** 15 *h.;* **d** 16 *h.*



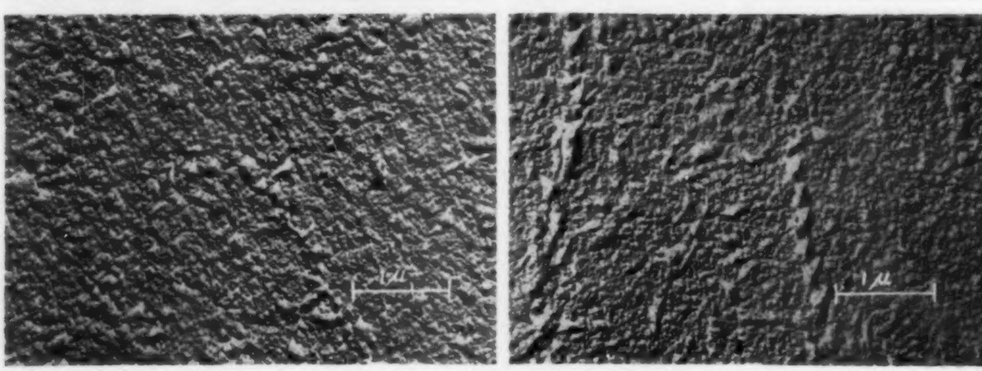
a b

10 *Structure of ferrite after hardening at a temperature of 97°C. for various periods:* **a** 15 *h.;* **b** 100 *h. Before hardening the specimen was subjected to plastic deformation* (ε = 20%)



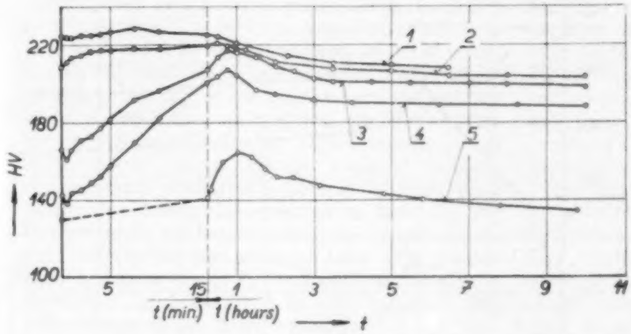
a b

12 *Structure of ferrite after hardening at a temperature of* 97°C. *for a period of* 15 *h., with preliminary hardening at a temperature of* 50°C. *for various periods:* **a** 1 *h.;* **b** 13 *h.*

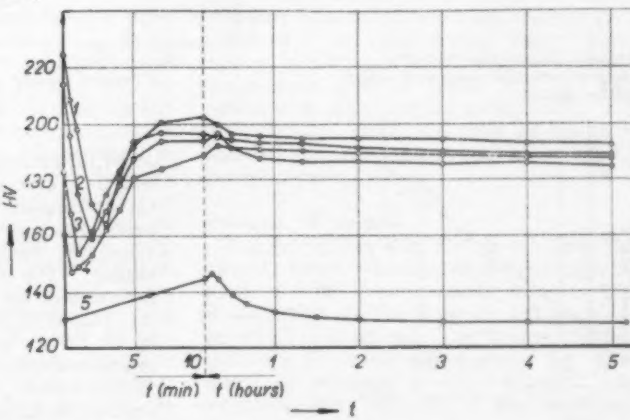


a b

14a and b *Structure of ferrite after hardening at a temperature of* 128°C. *for a period of* 9 *h. with preliminary hardening at a temperature of* 23°C. *for* 16 *h.*



11 *Time sequence of hardness during hardening at a temperature of 97°C. with preliminary hardening at a temperature of 50°C. for various periods: Curve 1 for 13 h.; curve 2 for 5 h.; curve 3 for 2 h.; curve 4 for 1 h. Curve 5 shows the hardness sequence during single-stage hardening at a temperature of 97°C.*



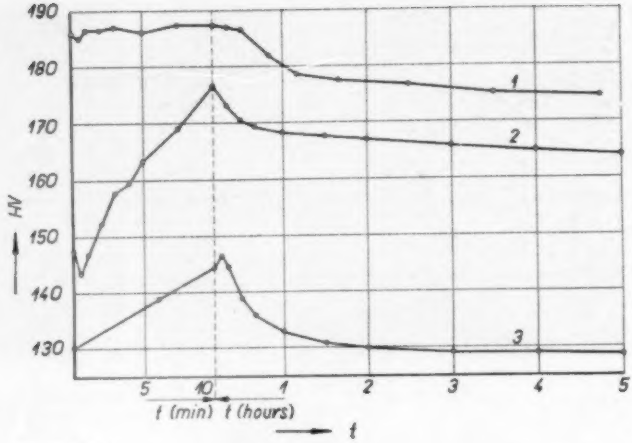
13 *Time sequence of hardness during hardening at a temperature of 128°C. with preliminary hardening at a temperature of 23°C. for various periods: Curve 1 for 168 h.; curve 2 for 100 h.; curve 3 for 34 h.; curve 4 for 16 h. Curve 5 shows the hardness sequence during single-stage hardening at a temperatue of 128°C.*

characteristic of the hardening sequence; with the increase in the hardening temperature the hardness rises more rapidly, but the maximum hardness diminishes. By hardening at a temperature of 23°C. a hardness maximum had still not been reached within the time period of the research.

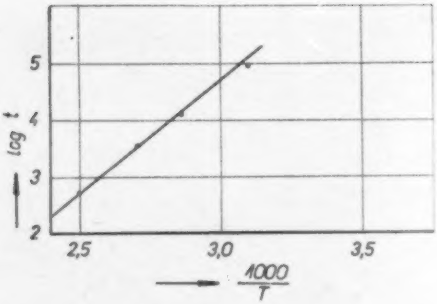
Precipitation at a temperature of 50°C. leads to a relatively considerable increase in the hardness of the ferrite; after reaching a maximum the hardness falls gradually. After 1 h. at this temperature, *i.e.* within a period when the hardness starts to increase, it is possible to distinguish a moderate wrinkling of the etched surface which is uniform within the limits of a single grain; the roughness of the surface differs for different grains (fig. 6*a*), which may be explained by their varying etchability in relation to the orientation of the individual grains with respect to the plane of the polished surface. Practically the same appearance is shown by the structure of a specimen after 16 hours' hardening at a temperature of 23°C. (fig. 5). After 6 h. at a temperature of 50°C. the character of the structure changes only slightly (fig. 6*b*); apart from the uniform wrinkling of the surface, signs are already apparent of a further roughening of the surface which is no longer uniform. This process advances with time, so that after 13 h. it is possible to distinguish small formations projecting above the matrix, which becomes gradually smoother (fig. 6*c*); the formations are manifested preferentially along a line which can probably be regarded as the boundary of the mosaic structure. A period of 13 h. corresponds to the maximum on the hardness curve (fig. 4). After 41 h., when the hardness has already fallen considerably, hundreds of small formations are apparent in the structure, which do not yet, however, have the character of clearly defined precipitates (fig. 6*d*).

A similar sequence is followed by changes in the structure at other precipitation temperatures also. At a temperature of 97°C. the rising sections of the hardness curve after 30 min. can again be associated with a uniform wrinkling of the surface (fig. 7*a*); the structure is similar to that in figs. 5 and 6*a*, which belong to specimens of about the same hardness. After 1 h. at a temperature of 97°C. the hardness maximum is reached, and the structure (fig. 7*b*) has a similar appearance to that

15 RIGHT *Time sequence of hardness during hardening at a temperature of 128°C. with preliminary hardening at a temperature of 77°C. for various periods: Curve 1 for 6 h.; curve 2 for 30 min. Curve 3 shows the hardness sequence during single-stage hardening at a temperature of 128°C.*



16 BELOW *Relationship between temperature and time required to attain maximum hardness*



in fig. 6*c* (500°C. for 13 h.—peak of the hardness curve), with the difference that it is radically coarser, and the number of formations projecting above the level of the matrix is less. After 15 h. at a temperature of 97°C. in the structure relatively coarse precipitates are clearly visible (fig. 7*c*), which often have a lamellar character and sometimes fairly irregular distribution (fig. 7*d*); to this state of the structure there already corresponds a hardness which is less than the hardness of the specimen after quenching.

A further advance in precipitation is evident after heating at a temperature of 128°C. for 9 h. The lamellar nature of the precipitates (fig. 8*a*) and their preferential precipitation in certain areas of the grain (fig. 8*b*) are clearly visible. After heating at a higher temperature (250°C. for 2 h.) the precipitates are very marked, and their number is relatively small (fig. 9).

A specimen of the steel which was subjected to a static tensile load ($\varepsilon = 20\%$), after quenching from a temperature of 700°C., increased its tensile strength as a result of the work hardening from 130 to 190 H_V. Precipitation at a temperature of 97°C. hardens such a specimen more rapidly and to a greater extent than specimens without such preliminary deformation (fig. 4). Within the period investigated, after attaining the maximum value, the hardness of the specimen fell only slightly. The wrinkling of the structure is more pronounced, the structure has an oriented nature (fig. 10*a*, 97°C. for 15 h.) and, by comparison with fig. 7*c* (97°C. for 15 h. without deformation), it can be established that no clearly marked precipitation takes place; in essence the state of the structure remains unchanged even after 100 h. at this temperature (fig. 10*b*).

The effect of a higher temperature on the hardness and microstructure of specimens of the steel hardened at a lower temperature was investigated in a further series of experiments. Quenched specimens were subjected to hardening at selected temperatures of 23, 50 and 77°C., and then exposed to the action of higher temperatures. In fig. 11 is shown the hardness sequence of the specimens whose preliminary hardening was carried out to varying degrees at a temperature of 50°C. (curve 1 for 13 h., curve 2 for 5 h., curve 3 for 2 h. and curve 4 for 1 h.), in relation to the time of heating at 97°C. The course of each of these curves was established by measurement of the hardness of the same specimen at normal temperature after the selected times of hardening at the given temperature.

For comparison, the hardness sequence is shown of a specimen hardened only at a temperature of 97°C. (curve 5). From the diagram it is evident that, by combined hardening at 50 and 97°C., higher hardness values are attained than by single-stage hardening at a temperature of 97°C. and that, after a moderate drop the hardness later decreases only slightly in relation to time.

Specimens which were preliminarily hardened at a temperature of 50°C. to a limited extent only (curves 3 and 4 in fig. 11) showed an initial sharp rise in hardness, reaching a maximum after about 45 min.; the rate of increase in hardness and the value of the maximum hardness are greater than for single-stage hardening at a temperature of 97°C. (curve 5), and the time required to reach this hardness is shorter. After a moderate drop in hardness during the course of the next 2 h. the further fall in hardness proceeds to a slight degree only. In the first part of curves 3 and 4 the gradual drop in hardness can be attributed to the reverse dissolution of nuclei, but in both instances this phenomenon is not clearly revealed.

The structure of the specimens (fig. 12*a*) after the combined hardening (50°C. for 1 h. + 97°C. for 15 h.) has essentially greater similarity with the structure of the specimen in fig. 7*c* (97°C. for 15 h.); the wrinkling of the surface is more marked than in fig. 6*a*, but clearly marked precipitation did not occur as in fig. 7*c* or *d*. The hardness of the specimen with the greater degree of preliminary hardening (curve 2, fig. 11) increases somewhat in the initial period and, after a gradual drop in hardness, further softening is again slight, with a greater hardness than that of curves 3 and 4. There is a similar hardness sequence of the specimen which during the preliminary hardening exceeded the earlier maximum hardness (50°C. for 13 h.—see fig. 4), wi h the distinction that the changes in hardness in the initial period are minimal (curve 1, fig. 11). The structure of the specimen preliminarily hardened to maximum hardness (fig. 12*b*—50°C. for 13 h. + 97°C. for 15 h.) is similar to that in fig. 12*a*; even here precipitates of characteristic shapes are not apparent. The wrinkling of the surface, which is very markedly manifested, can scarcely be qualified as a manifestation of the precipitation of a new, incoherent phase.

A change in the temperature conditions of the combined hardening somewhat alters the sequence of the curves, but does not alter the general nature of the phenomenon investigated. In fig. 13, in the same way as in fig. 11, is shown the hardness sequence of specimens of the steel after combined hardening (23°C. + 128°C., curves 1–4), and curve 5 shows again the hardness of a specimen which was hardened simply at a temperature of 128°C. As distinct from fig. 11, for all the preliminarily hardened specimens there is an evident drop in hardness in the initial period, which is all the greater the higher was the level reached by the preliminary hardening; it may be attributed to the reverse dissolution of nuclei in the alpha solid solution. After the transitional drop in hardness, there follows a marked and rapid rise in hardness, and after a moderate fall during the first hour the hardness subsequently falls only slightly. The resultant hardness is again substantially higher than after simple hardening at a temperature of 128°C.; the differences in hardness between specimens with various states of preliminary hardening are small, and there is no logical relationship between the degree of preliminary hardening and the final hardness after combined hardening.

The microstructure in fig. 14*a* (23°C. for 16 h. + 128°C. for 9 h.) has a similar appearance to that in fig. 12*a*, but in some areas there are apparent fine lamellar precipitates, which mutually form oriented networks (fig. 14*b*). By comparison with single-stage hardening at a higher temperature (figs. 8*a* and *b*—128°C. for 9 h.), the number of lamellar precipitates is very small and they are finer.

Basically, the same results follow from the further investigation of combined hardening at temperatures of 77° + 128°C. (fig. 15). By comparison with the preceding instance, after reaching the maximum value, the drop in hardness and likewise the subsequent gradual softening are somewhat more marked.

Analysis of results

The low-carbon steel with a very low nitrogen content, which was selected for the experiments, makes possible the assumption that the sequence of hardening will be decisively influenced by the precipitation of carbon. The hardness curves after hardening at various temperatures (fig. 4) are in agreement with the generally accepted opinions concerning the sequence of the breakdown of a solid solution through precipitation. The sequence of this phase transformation through diffusion must be controlled by the diffusion of carbon in alpha iron.

The time t required to reach a certain state of precipitation is given by the equation:[13]

$$t = x^2/2D \qquad (1)$$

where x is the mean length of the diffusion path and D is the diffusion coefficient, for which the known relationship is valid:

$$D = D_0\, e^{-Q/RT} \qquad (2)$$

in which Q denotes the activation energy of the diffusion.

After insertion into equation (1) and rearrangement we have the relationship

$$\log t = K_1 + K_2\,(1{,}000/T) \qquad (3)$$

in which

$$K_1 = \log x^2/2D_0$$
$$K_2 = Q \log e/1{,}000\, R$$

On the assumption that analogous points on the curves of the changes in hardness in relation to

the time of hardening at various temperatures correspond to similar states of precipitation, in accordance with equation (3) in a graph with coordinate axes log t and $1{,}000/T$ the corresponding points must lie on a straight line; if we transfer to this system the plots of the maxima of the hardness curves (fig. 4), with them we can with good approximation interpolate a straight line (fig. 16). From the gradient of this straight line it is possible to determine the activation energy of the precipitation process under investigation, which amounts to $Q = 18{,}000$ cal./mol.

This value is relatively close to the activation energy of the diffusion of carbon, which is 20,100 gal./mol. according to Wert.[16] This relatively good agreement confirms that carbon diffusion is the process which controls the hardening of the super-saturated solution of alpha iron.

From the micrographs of the structures it follows that it is very difficult to establish when true precipitation takes place, *i.e.* when a crystallographically defined phase is formed, which is separated from the solid solution by inter-phase boundaries. Examples of a structure of this type are figs. 7*c*, 7*d*, 8*a*, 8*b* and 9, from which it is evident that the less easily etched formations suggest regular, geometrical bodies with an ordered orientation within a single grain. These states collectively correspond to the very low hardness values on the descending sections of the hardness curves.

The structures of the specimens on the rising section of the hardness curves (figs. 5, 6*a*, 6*b* and 7*a*) after etching have a wrinkled surface, which is a result of the varying degrees of etchability of individual areas of the surface of the specimen; the phenomenon may be explained by the inhomogeneity of the alpha solid solution. The structure of specimens with a hardness approximately corresponding to the peak of the curve and the section closely following its peak represents to some extent a transition between the two types of structure described above, but here it is still impossible to point to clearly marked precipitation (figs. 6*c*, *d* and 7*b*). It is probable that it is a matter of coherent precipitates, *i.e.* formations which within a certain restricted zone have their own lattice, but from it there is continuous transition without an inter-phase boundary into the lattice of the solid solution.

Analysis of the structures revealed by the electron microscope therefore indicates that the hardness maximum belongs to the pre-precipitation state of the inhomogeneous solid solution or to the state of coherent precipitates in this solid solution; progressive precipitation of carbide formations leads to a reduction in hardness. The finer is the structure, *i.e.* the greater the number of coherent zones, the higher will be the hardness of the specimen. At a higher hardening temperature there is a correspondingly greater coarseness of the structure, and the continuous transition from an inhomogeneous solid solution by way of coherent precipitates to incoherent precipitates is more rapid.

Apart from the carbon diffusion which has an influence on the formation of nuclei and especially on their growth, an important effect on the sequence of precipitation is exerted by the state of the structure, which in essence determines the nucleation potential of the matrix. The important influence of defects in the structure during the breakdown of the cementite and the precipitation of graphite, and their static and dynamic character, we have already shown in an earlier work.[17] For the instance of the precipitation of ε-carbide it may rightly be assumed that the nucleation potential of the matrix, determined by the quantity and type of the structural defects, does not change greatly within the limits of the precipitation temperatures investigated (23–128°C.). Under constant nucleation conditions, therefore, the sequence of the precipitation at various temperatures is in essence dependent only on the rate of carbon diffusion, and the curves have a similar sequence. At higher hardening temperatures the rate of carbon diffusion is higher, the whole process advances more rapidly, and the changes in hardness take place within a shorter time; the hardness maxima are lowered, because at higher temperatures the critical dimension of a nucleus is greater, a smaller number of nuclei are capable of growth, and therefore a smaller number of coherent zones are formed, and in the final stage of precipitation a smaller number of precipitates also. In all stages of the process, therefore, at higher temperatures the structure has a coarser character.

The nucleation potential of a super-saturated alpha solid solution can be changed, for instance, by cold working; the density of the dislocations increases up to 10^{12}/cm.^{2} from a value of 10^8/cm.^{2} in the annealed state.[18] As a result of the more favourable conditions afforded by the high concentration of structural defects, very rapidly a large number of nuclei are formed, which change into coherent zones. The precipitation process follows its course at a given temperature substantially more rapidly than for unworked material; the hardness increases more rapidly and to a greater extent (fig. 4). A similar favourable effect of cold working on the nucleation of graphite and on the course of graphitization of sorbitic cementite has already been shown earlier.[17] In addition it was shown in the same work that structural defects formed in the matrix as a result of cold working, have a stable nature in their nucleation action, and the nucleation potential of the matrix cannot be influenced by

regulation of the temperature conditions of the precipitation process.

The increased stability of the hardness and of the structure during the hardening of low-carbon steel after mechanical working is manifested by the high stability of the coherent zones forming in the areas of the structural defects. The interaction of the dislocations, where their concentration is high, creates in the matrix favourable conditions for the formation of a large number of nuclei of supercritical size, and the coherent zones forming on them are relatively stable against dissolution and against growth. A lack of zones which readily dissolve restricts the possibility of the growth of the coherent zones of supercritical size, which would probably transform into incoherent precipitates on attaining a certain minimum size. Therefore the hardness scarcely falls, and the structure is unchanged. The relative stabilization of a certain metastable state is in this instance the result of control of the nucleation potential by the fact that new stable defects were introduced into the structure by mechanical working.

From the results presented it follows that a certain stabilization of the hardness and of the structure can be achieved by combined hardening also. The process of hardening is in this instance more complicated than during single-stage hardening, and its individual states are most clearly marked under conditions of combined hardening at 23°C. + 128°C. (fig. 13).

By hardening in the first stage at 23°C. a pre-precipitation state is formed, corresponding to the length of time of hardening. After heating at the temperature of the second stage, 128°C., there ensues a sudden drop in hardness, because the coherent zones are partially dissolved in the alpha solid solution. The drop in hardness is greater, and the absolute values of the hardness minima are higher with a greater degree of precipitation in the first stage, and the time required to reach the minimum, on the other hand, increases; the minimum hardness always has a higher value than the hardness of the steel after quenching.

Let us assume that, in accordance with the *theory of fluctuations*, the extent of the coherent zones for each stage will be given by a *frequency curve*. At a low degree of hardening in the first stage, after heating at the temperature of the second stage there is very rapid dissolution of the formations which are smaller in size than the critical dimension of a nucleus for the temperature of the second stage. The first impulse towards dissolution can be increased solubility of the matrix in the vicinity of a coherent zone, which allows the passage of carbon atoms into the solid solution; by this process there is a further reduction in the dimension of the coherent zone and thereby in its stability. The greater drop in hardness after a longer period of hardening in the first stage means that there was a growth in these zones which dissolve by comparison with the state where the extent of the hardening in the first stage was small. The higher value of the hardness minimum indicates that after a greater degree of hardening in the first stage there was a growth in the number of formations which are not dissolved at the higher temperature. During the course of hardening in the first stage, therefore, the hardness rises both through the gradual formation of coherent zones and through their growth. Undissolved or only partially dissolved coherent zones become very effective nuclei for hardening in the second stage, and therefore the hardness increases very sharply, more rapidly than during single-stage hardening.

After attaining a not-too-pronounced hardness maximum there follows a period of imperceptible reduction in hardness after a moderate drop. This stabilization of the hardness is evidently the result of the fact that in the initial period the subcritical coherent zones are eliminated by dissolution, and a certain homogenization of the size of these formations takes place. By this means there is a substantial reduction in the possibility of dissolution, and also of growth, of the coherent zones in the further period of hardening, so once again the hardness and the structure become relatively stable. As a result of the favourable nucleation state of the matrix after the formation of a large number of nuclei in the first stage of hardening a higher number of coherent zones remain preserved, the structure is finer, and the hardness is greater than by means of single-stage hardening at the temperature of the second stage. The sequence of hardening and the properties after the period of the process investigated (hardness and structure) are dependent on the conditions of the first stage of hardening (temperature, time) and the temperature of the second stage of hardening.

Conclusions

Hardening of low-carbon steel with a low nitrogen content proceeds through the formation of a carbide phase. The activation energy of the precipitation process established is very close to the activation energy of carbon diffusion in alpha iron, which shows that carbon diffusion is the process which controls the sequence of hardening of the steel under investigation.

By means of the electron microscope it is possible to follow the changes in the structure of the solid solution of alpha iron in relation to the time and conditions of hardening. But determination of the start of the actual precipitation, *i.e.* the precipitation of incoherent formations, is difficult. Clearly

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Physical Society Exhibition

The Physical Society's annual exhibition of scientific instruments and apparatus is one of the activities which has been taken over by the amalgamated Institute of Physics and Physical Society. It took place as usual in the Royal Horticultural Society's Halls in London last month and was the 45th in the series. Its chief object, it will be remembered, is to reveal new ideas, principles and inventions and subsequently to demonstrate their realization in the form of usable equipment. Although the tunnel effect of semiconductor diodes was first described by a Japanese scientist some years ago, this year, for the first time, five stands of the exhibition featured this new type of transistor. Another effect which had been noted many years ago was that an electron beam in a vacuum exerts an etching effect on metals: the Research Department of Associated Electrical Industries (Manchester) Ltd. make use of this effect for the production of etched patterns on metallic surfaces. Another idea of considerable age has recently been twisted about: semiconductor thermocouples are now used—as exemplified by the exhibits of Joseph Lucas Ltd. and the Research Laboratories of the General Electric Co. Ltd.—as refrigerators. The idea of using magnetic materials as storage devices for computors has been developed and four exhibitors are featuring thin magnetic films with storage capacities of the order of 2,500 ' bits ' with switching times less than 10^{-9} sec. As usual, the breadth of the research effort of Government Establishments is most impressive in both theoretical and experimental fields. (Ministry of Aviation, Department of Scientific & Industrial Research, Admiralty Research Establishments, United Kingdom Atomic Energy Authority.) At the same time it is now becoming noticeable that all firms, however small, appear to have reasonably effective research groups, capable of some original work. The following notes give details of some items of metallurgical interest

THE TELE-REMOSCOPE by Cawkell Research & Electronics Ltd. is a new type of television picture monitor which allows the user to produce on the cathode-ray tube screen a still image at any instant during a changing picture. The resulting image may be examined for long periods. This monitor can be applied to the inspection for moving or momentarily stationary objects in automatic production lines. By ' freezing ' the image the inspection time can be greatly increased over that normally available. This, and other features, were demonstrated on a prototype.

The exhibition provided a good opportunity for the various research associations to propagate their results, and this year the DSIR stand housed some two dozen exhibits. Here one learnt that the National Engineering Laboratory has constructed an apparatus which will generate pressures in excess of 100,000 atmospheres in conjunction with temperatures of over 2,000°C. The formation of new materials of technical interest, such as the synthetic diamond and cubic boron nitride, will be studied with this apparatus.

High-temperature ceramics

New high-temperature ceramics to supersede metals and traditional ceramics are being produced on a pilot-plant scale at the laboratories of the British Ceramic Research Association. Among those exhibited were boron nitride, a low-friction, machinable electrical insulator, silicon nitride, a strong, chemically inert material, and self-bonded silicon carbide, a dense material of high strength and erosion resistance at high temperatures.

The British Glass Industry Research Association exhibited an automatic thermal expansion apparatus, which, although initially designed for measurement of the thermal expansion of glasses up to the annealing point, is eminently suitable for dilatometric measurements of other materials. The instrument,

which has a built-in calibration unit and accommodates a 4-in. sample length, can be semi- or fully-automatic as required, and in its present form may be used for annealing and heating cycles up to 800°C., but this temperature can be extended.

A carrier-gas method of hydrogen determination has been developed by the British Welding Research Association. A determination takes about 15 min. Argon is passed over a sample of steel, which is heated in a furnace. The hydrogen evolved is swept along in the argon stream to a hydrogen sensitive electric conductivity cell, the output of which is plotted on a conventional chart recorder. Ingress of air during sample loading is avoided by a special loading lock and a special burette is used for injecting small measured volumes of hydrogen gas for calibration. The estimated precision of the method, using a 5-g. steel sample containing 1 ml. hydrogen/100 g. steel, is $\pm 0 \cdot 04$ ml./100 g.

The exhibit of the British Steel Castings Research Association was a radioactive tracer technique for investigating the origin of inclusions in steel castings. In this technique molten steel flows over clay-bonded sand compacts in which the silica grains have been coated with activated ceria. Autoradiography is then employed to study the inclusions in the steel casting and elucidate the mechanism of their formation from interaction between slag, steel, etc., and the sand.

Continuous control methods

Today it is the continuous high-rate production lines which are demanding additional measuring and control instruments. An urgent demand arises in the production of steel strip. Here, the continuous measurement of the reduction in gauge in the temper rolling of strip is a demand still unfulfilled. A new approach to the problem was exhibited by the British Iron and Steel Research Association. In the prototype instrument two photocells are sighted on two illuminated spots on the strip surface, the two spots being a fixed distance apart and lying along the passline. Small irregularities in the light reflected from the strip surface give rise to a pattern in the output of the first photocell. After an interval of time, depending on the distance apart of the two photocells and the strip speed, another similar pattern appears in the second photocell.

Thus, if the output of the first photocell signal is delayed (*e.g.* on a tape recorder) and then added to a second photocell signal, a maximum combined output will be obtained when the delay time introduced is equal to the time for passage of a point on the strip surface between the two illuminated spots. From this delay time and the known distance between the illuminated spots the strip speed can be readily calculated.

Improved steel-strip production is the objective of another BISRA development, which, when perfected, will enable tinplate and cold-rolled strip to be inspected in a high-speed line for surface defects. The prototype instrument has reached the 'field trial' stage.

New thermocouple elements

New thermocouple elements for measuring temperatures in the range 1,600–2,000°C. were exhibited by the Morgan Crucible Co. Ltd. Some of the materials, being semi-conductors, provide a very high voltage output compared with metals. For example, a silicon carbide-graphite couple can be made to give 0·2 V. at 1,000°C. Modification of the properties of silicon carbide and graphite, by controlled adjustment of composition, allows couples to be made from two types of silicon carbide, or two types of graphite. Alternatively, any of these materials can be run against a refractory metal such as tungsten or rhenium.

Also among the exhibits of the Morgan Crucible Co. was a technique for hot-stage time-lapse ciné photomicrography, which allows one to follow under a microscope the slow changes in crystal form or other visible qualities of solid materials. An electrically operated camera is used to take a series of single shots with variable exposures at intervals of 1 min. The resulting ciné film can be used for quantitative measurements.

Metallurgical investigation

New and improved techniques for metallurgical investigation figured prominently in the exhibition. Some of these require expenditure on a scale that can only be met by a large organization and others only a modest outlay. Cooke, Troughton & Simms Ltd. exhibited a stereoscopic microscope of different design from the conventional form of stereoscopic microscope. Instead of having twin objectives inclined inwards so that their optic axes meet in the plane of the specimen, it uses one objective, and an optical system in the body of the microscope passes left- and right-handed images to the eyepieces to produce the sensation of depth. Thus the same effect is produced as with the more usual type of stereoscopic microscope but with the advantages of a constant working distance and a flat field of view.

A scanning electron-probe X-ray microanalyser, now in production, was exhibited by the Cambridge Instrument Co. Ltd. The technique of X-ray microanalysis consists of focusing an electron beam, normally accelerated from 0 to 50 kV., on to a region of the specimen surface approximately 1 in. dia. and rather less in depth and analysing spectroscopically the resulting X-rays excited from the specimen. So far, the technique has been used almost exclusively for research work on metallurgical

problems, although it also has applications in biological and other fields.

As well as static probe point-by-point quantitative microanalyses, the Cambridge instrument can also be used for scanning the surface of the specimen over an area up to ½ mm. by ½ mm. The characteristic X-rays from any single element (with atomic number 12 or greater) can be selected by a crystal spectrometer and detected by a proportional counter. The signal from the counter modulates the brightness of a beam scanning a cathode-ray tube, in synchronism with the probe scanning the specimen, and forms an image corresponding with the area scanned and showing semi-quantitatively the distribution of the chosen element.

The accuracy for quantitative analysis is to $\pm 1\%$ which reduces, under favourable conditions, to $0 \cdot 1\%$. The minimum concentration that can be detected under good conditions is $0 \cdot 1\%$.

Among the Post Office Engineering Department exhibits was an inexpensive technique for studying plastic flow which can be applied to the study of a wide range of metallurgical problems such as plastic flow in forming and extrusion. Devised originally for investigating contact deformation, it enables the effects of friction and rate of deformation, and of shape and internal constraint, to be readily examined. The technique employs two-colour laminated test-bodies made from ' Plasticine '; the laminations can be oriented in any direction.

New vibratory and electrolytic polishing equipment was shown by Nash & Thompson Ltd. The vibratory polisher can polish in one loading any number of specimens from two to 18 and will operate with no attention. Thus a batch of specimens can be polished overnight. A finish suitable for electron-microscope examination can be obtained. The electrolytic polisher consists of a ' swirl head ' assembly into which the specimen is mounted and subjected to an applied potential while the surface is in contact with a continuous flow of electrolyte.

Precipitation hardening

concluded from page 73

defined precipitates were found to correspond to the descending sections of the hardness curves; the rising section of the curves and the vicinity of the maximum coincide with the pre-precipitation state of the inhomogeneous solid solution or the state of coherent precipitates.

After cold working and after combined hardening at two different temperatures higher hardness values are attained, and there was found to be considerable stabilization of the structure and hardness by comparison with single-stage hardening at the given higher temperature and hardening of unworked steel. Both phenomena may be explained by the considerable nucleation potential of the alpha solid solution.

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Creep rupture testing with ½°C. temperature control

Electric furnace zone control to within ½°C. or closer has been achieved at a new creep/rupture testing installation at Colvilles Ltd., Research Department, Mossend, near Glasgow. This is believed to represent finer control of temperature than has been obtained hitherto in this type of work.

The first of six new electric resistance furnaces, which will be used to collect creep/rupture data on various steels which Colvilles produce, is now fully working under control of instrumentation designed to keep the temperature of specimens and furnace zones within very close limits.

The six furnaces will each operate at a different temperature (± 10°C.) from 400 to 550°C. The temperature control is well within the tolerances laid down by the appropriate British Standards governing this type of test.

Each furnace is itself divided into six zones containing thermocouples; thermocouples are also placed adjacent to the specimens. Using Colvilles own switchgear panel, measurements of these points are taken in sequence, thermocouples being connected for a few seconds each. A continuous record is given on a Honeywell Controls Ltd. ' ElectroniK ' strip chart recorder-controller; the ink trace is continuous since there is an ' overlap ' in the switchgear which provides the average of the temperature of two zones when switching over from one zone to another.

The instrument acts as a controller in conjunction with the switchgear panel and experience with the first furnace shows a maximum variation of ½°C. per zone and 1½°C. over the whole furnace under the worst combination of errors.

An additional safeguard covering the whole circuit is provided by means of a Honeywell ' Protect-O-Vane ' excess temperature safety cut-off. The creep/rupture tests are carried out over extended periods, and, although low temperature would be tolerated, a single instance of high temperature would ruin the test and waste much time. This overall protection in the case of each furnace has therefore been thought prudent.

Metallurgy in nuclear power technology

6. Fuel element technology

J. C. WRIGHT, B.Sc., Ph.D., A.I.M.

The metallurgy of nuclear power materials is developing on such a wide front and so rapidly that it is difficult for the non-specialist metallurgist to keep abreast with its scope. Dr. Wright, Reader in Industrial Metallurgy, College of Advanced Technology, Birmingham, outlines the subject in a series of articles which are appearing monthly in this journal

Before reviewing present and possible future trends in the development of reactor fuels it is advisable to consider the ideal properties required of a reactor fuel and the special difficulties met with in the working of a reactor. Ideally, a reactor fuel should have the following properties well developed:

(1) During fission it should produce sufficient neutrons, in excess of those which will be absorbed uselessly, to continue the fission process with a margin for control purposes. In other words, the fuel should have good neutron economy.

(2) The fission process produces heat throughout the whole volume of a fuel element and, to be useful, this heat must be transferred to a medium which is capable, after a further heat exchange, of passing the energy to a turbine generator set. The fuel element must therefore contribute to good heat transfer without reacting with the coolant.

(3) The fuel element must be able to prevent release of fission products, which could be very radioactive, to the coolant circuit.

(4) Probably the most difficult property to achieve satisfactorily at present, and the one associated with the physical metallurgy of a fuel element to the greatest extent, is that the fuel should be capable of a high burn-up without mechanical failure.

Most of the special difficulties met with in the reactor directly concern this last property. Summarizing the difficulties, there is the possibility of damage due to: (1) Phase changes and transformation damage; (2) surface roughening and growth due to thermal cycling; (3) wrinkling and growth due to irradiation effects; (4) swelling due to irradiation effects above 450°C.; (5) changes in mechanical properties; and (6) changes in physical properties.

Although the above factors mainly concern the control of mechanical failure, the other properties of a reactor fuel may well be affected when attempts are made to overcome mechanical failures. For instance, alloying elements must be chosen with due regard to neutron absorption cross-section or the neutron economy will suffer.

Control of distortion

Since a large proportion of the damage to uranium under reactor conditions is due to the marked anisotropy of the α-phase and poor high-temperature strength, two major lines of development have been pursued. In the first, the effects of the anisotropy of α-uranium have been minimized by achieving fine and randomly oriented grains of uranium by methods described in previous sections. Correctly applied, these techniques are capable of controlling growth and wrinkling of uranium fuel bars sufficiently well for present reactor purposes. The higher temperature problem of swelling is becoming more and more important as reactor temperatures rise and is not controllable, in pure uranium at least, to anything like the same extent as growth and wrinkling. The creep strength of a fuel element must be high to resist the expansion of gas bubbles which create swelling. If the phase possessing a high creep strength was also fully isotropic, this would assist in controlling growth and wrinkling. Evidently, it is desirable to develop a strong isotropic fuel.

An example of such a development is provided by work on a uranium/12% molybdenum alloy which is single phase (γ) over a wide composition

and temperature range and relatively strong. Such an alloy has been shown to resist swelling better than α or $\alpha + \beta$ uranium. The phase is isotropic and can be retained to lower temperatures than the γ-phase of pure uranium. Its mechanical properties (Table 14) are far superior to those of α-uranium under the same conditions. The phase diagram of the U-Mo system is shown in Fig. 37. Unfortunately, molybdenum has a moderately high neutron-capture cross-section, particularly for thermal neutrons, 2·4 barns. This limits the amount of molybdenum which can safely be introduced. It has been suggested, however, that the γ-phase in U-Mo alloys can be held to a lower temperature under irradiation conditions than under equilibrium conditions when the alloy would normally be $\alpha + \gamma$ uranium/molybdenum. This would allow the use of rather less molybdenum to achieve a satisfactory phase under reactor conditions. Similar depression of transformations under radiation conditions has been noted in uranium/niobium alloys. The depression of the transformation is probably due to the increased ease of forming a stable nucleus of the higher-temperature phase under the disturbing influence of irradiation.

Other alloys with extended γ-phase fields are provided by the systems U-Zr; U-Nb; U-Ti; but the latter introduces a rather high neutron-absorption coefficient.

If the gases causing swelling could be removed from the matrix before they had chance to expand into bubbles, swelling would be controlled at source. It should be possible to remove the gas atoms from the fuel before they can nucleate and form bubbles if the distance the atoms move to find a free surface is small. The obvious way to provide easy access to a free surface is to make the uranium in the form of a powder compact with a large amount of inter-connecting porosity. It is, in fact, possible to make a compact of this type with uranium powder of about 10 μ particle size, but the swelling observed in such compacts under irradiation has not been significantly different from that in solid material and the amount of gas released was very small. Apparently, the gas is able to nucleate bubbles even before it has diffused to the surface of a particle of powder in the compact. There may still be a possibility, at temperatures higher than are normal for fuel elements at present, for the gases to escape before forming bubbles or it may eventually be possible to use even finer powders. For the moment, however, this approach to the problem is not a satisfactory one.

TABLE 14. *Some properties of heat-treated U/12% Mo alloy (test temperature = 600°F.) (H. A. Saller and F. A. Rough, Int. Conf. P.U.A.E., 1955, P/558)*

Prior heat-treatment	Tensile strength, lb./sq. in.	Elongation, %
24 h. 900°C. W.Q.	108,000	3·6
24 h. 900°C. W.Q.; aged 3 h. at 500°C.	104,000	5·8
24 h. 900°C. W.Q.; aged 40 h. at 500°C.	118,000	5·3

If, for the present, swelling cannot be eliminated at source, it becomes necessary to use various methods of restraint so that the swelling pressure created by the gas is opposed by strong hydrostatic pressure in an attempt to prevent the density decreasing. The difficulty is one involving a compromise between can strength and neutron economy. In addition, a further heat-transfer resistance is created. With a reasonable compromise, this method increases the available percentage burn-up and safeguards against premature breaking away of fuel particles and escape of fission products.

Internal restraint

Instead of having a complete can as a restraining device, it is feasible to restrain small amounts of fissionable material by dispersing it within a strong, non-fissionable matrix. The latter then prevents macroscopic swelling to a large extent and may allow higher amounts of burn-up without failure. The disadvantage of this method is that dilution reduces the active concentration of the fuel for a given volume and also introduces another neutron absorber. The diluent must therefore have a low neutron capture cross-section and the volume restrictions on reactor size may enforce the use of an enriched uranium fuel or plutonium.

Another type of internal restraint may be provided for uranium matrices by creating within them a rigid network. It is possible to produce, for instance, a network of uranium carbide within uranium, the whole having a density of about 15·5. This system is likely to withstand swelling and thermal cycling across the α/β transformation without suffering the distortions produced in pure uranium.

Raising the creep strength of a fuel element by alloying additions has been mentioned previously. In general, the amount of alloying element addition permissible without too much effect on the neutron economy is not sufficient to enhance the strength of the fuel element enough to withstand radiation damage. However, if a concentration of alloying element is made in the outer regions of a fuel element, and pure uranium or a uranium-rich phase contained within it, it might be possible to get adequate properties from the outer material to support the whole without unduly affecting the neutron economy. A powder metallurgy method could be

used to achieve the non-uniform distribution of the alloying element.

The powder method is very attractive in several ways, particularly when combining a dense metal, like uranium, with a second material which may be very light. Melting and casting would normally lead to segregation, which would require considerable efforts to approach homogeneity. Powder metallurgy offers the advantages that materials of widely differing densities can be intimately and uniformly mixed and phases which would react in the liquid state can be kept stable and separate because no melting is necessary. The most difficult aspect of producing satisfactory powder compacts is in choosing the correct particle sizes and shapes for the various components of the system, and all materials used usually have to be very pure. The maintenance of this high purity may require that all operations are carried out in an inert atmosphere, for instance. Finely divided Zr, Ti, U, and UC and UO_2 all come into this category.

The materials and methods which may be used in producing the various types of internally restrained fuel elements are worth studying. As an example of the dispersion of enriched uranium in a pure metal phase, it is possible to design a feasible fuel element of aluminium containing up to about 25 wt.% uranium. Alloys of uranium and aluminium with 14–25 wt.% uranium consist of crystals of UAl_4 in a matrix of Al/UAl_4 eutectic. It is possible to exceed 25 wt.% uranium in aluminium, but precautions must be taken to avoid segregation.

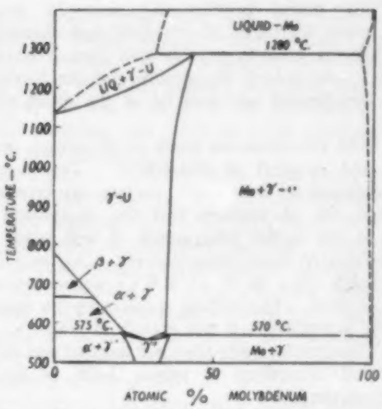
Two production schemes for this system, one employing fusion and the other powder methods, might be outlined as follows:

Fusion method

1. Pure aluminium and enriched uranium is melted in a graphite crucible in an open atmosphere using H.F. induction heating.

TABLE 15 *Reactions between U and Al* (*R. Kiessling, Int. Conf. P.U.A.E.*, 1955, *P*/786)

Temp. °C.	Time until reaction was noted	Resultant phases (in order of decreasing amounts)
250	~240 h.	UAl_2
300	24 h.	UAl_2, UAl_4, UAl_3 (trace)
350	6 h.	UAl_3, UAl_4, UAl_3
400	~1 h.	UAl_3, UAl_2, UAl_4
450	—	UAl_3, UAl_4, UAl_2
500	Immediate	UAl_3, UAl_4, UAl_2
600	Immediate	UAl_3, UAl_4, UAl_2



37 *The uranium/molybdenum equilibrium diagram*

2. The combined melt is stirred, degassed and cast into a graphite slab mould. Casting temperature 700 to 1,000°C., depending on composition.

3. The slab is fabricated by hot forging, hot rolling and cold rolling to about ¼-in. strip. Hot working temperatures depend on the alloy composition, but are of the order of 600°C.

4. Blanks are then punched from the strip.

5. Each blank is then fitted into a frame of aluminium and cover plates of aluminium are placed top and bottom.

6. The fuel plate assembly is then heated to 590°C. and hot rolled to give a reduction of over 80%.

7. The plates are then annealed at 610°C., after coating them with a slurry of alcohol and brazing flux. This treatment is followed by a thorough wash which removes the eutectic mixture of sodium, lithium and potassium fluorides and chlorides, and then by dipping in a dilute nitric/H.F. acid bath, washing and drying. The flux anneal inhibits blistering at a later stage.

8. The plates are then cold rolled to a final thickness of 0·06 in. and annealed at 590°C.

9. Non-destructive testing inspection follows to ensure straightness, freedom from defects and to locate the core.

10. The plates are then sheared to final size, assembled into fuel element packs with accurate spacings and brazed into the assembly framework using a silicon aluminium brazing alloy.

Powder method

This method is more commonly used for relatively high uranium-content alloys, of the order of 25 wt.% uranium upwards.

1. The initial powder is made by melting appropriate mixtures of uranium and aluminium, allowing the melt to solidify and then crushing to powder. Atomized aluminium powder forms the other component and will be of the order of 325 mesh.

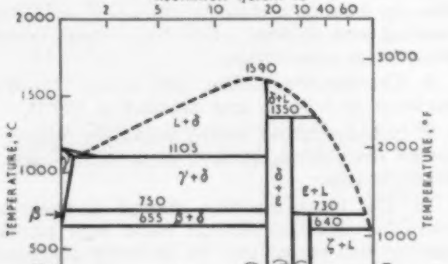
2. Cold compacts are made at 40 ton/sq. in. and these are sintered at 450–600°C. The sintering time depends on how much reaction can be allowed between the aluminium and the compound. If UAl_2 is the initial compound, it will react with aluminium to form progressively UAl_3 and UAl_4 (see Table 15); 96 h. at 600°C. will convert the UAl_2 to UAl_4. Hot rolling accelerates the reaction, but the particle size is not greatly altered.

3. The compacts are then hot worked and further processed according to stages 3–10 as for the fusion method.

In the powder method of approach, the size and dispersion of the particles of compound can be controlled to a large extent by selecting suitable powder sizes to start with. The fusion method relies to a greater extent on heat-treatment procedure for controlling the microstructure. The cast material solidifies with a fine, needle-like precipitate of UAl_4 dispersed in an aluminium matrix. After hot-rolling at 430°C., a matted needle structure of the UAl_4 in aluminium is produced, but on annealing for a time about 1 h. at a temperature between 400 and 600°C. the needle structure agglomerates into a satisfactory coarse dispersion.

Other dispersion systems of interest include uranium in thorium; in zirconium; and in beryllium. Table 16 lists properties of some possible fissile dispersions.

A properly designed fuel element of the dispersion/matrix type calls for a uniform dispersion of



38 *Constitutional diagram for the uranium-aluminium system* (USAEC *Report, BMI*/1,000)

the fissile phase in the non-fissile matrix. The particle size, shape and distribution should be such that the region of damage due to irradiation of one particle does not overlap the damage zone of the next. In this way, at all times up to processing of the partly burnt-up fuel element, there will be a continuous network of relatively undamaged matrix. The larger the particle size, the less likely are the irradiation damaged zones likely to touch, but in order to control the shape and dispersion characteristics, particle sizes are maintained at about 100 microns. The fissile material should be as rich in fissile uranium as possible and sufficiently strong to retain its size and shape during fabrication. Interaction with the matrix should be at a minimum and the matrix must have adequate mechanical properties, even under irradiation, to hold the macroscopic dimensions of the fuel element constant. Both the dispersed phase and the matrix should have low neutron-capture cross-sections. Additionally, the matrix should have a good thermal conductivity, low coefficient of thermal expansion and metallurgical compatibility with any cladding which is to be used.

Obviously, this is a rigorous scheme of properties and to superimpose on it the added difficulties of melting and casting would reduce the number of possible systems to a very low level. Fortunately, many more systems become possible if powder-metallurgical methods are introduced.

A higher concentration of uranium in an aluminium matrix is attainable when the uranium is present as UO_2 rather than as UAl_4. For instance, with 40% of uranium in aluminium, much of the aluminium is combined as UAl_4, leaving little for the matrix (fig. 38). The alloy is therefore brittle. When UO_2 is present in an aluminium matrix, no reaction takes place and a weight concentration of over 50% uranium is possible. It is possible to introduce the oxide into the aluminium as U_3O_8 and this is at least partially reduced to UO_2 by the aluminium during processing. Such a system is, of course, a cermet and aims at overcoming two of the major difficulties experienced with the pure ceramic fuels discussed in the next section. The ceramics have some very attractive properties but they are very brittle and subject to failure by thermal shock and generally they have poor thermal conductivities. The cermet overcomes these difficulties to a large extent, since the metal matrix can have a high thermal conductivity and act as the main heat-transfer medium. The matrix may also be ductile and contain without difficulty the brittle ceramic phase.

In order to achieve a high concentration of fissile atoms the cermet must be rich in the ceramic part, but if it is too rich the cermet will be difficult to fabricate. Powder-metallurgy methods may be

used to overcome this difficulty if necessary, but it is sometimes preferable to use normal metal fabrication techniques. The temperature of working must be sufficiently high to render the matrix ductile so that the brittle phase is not damaged.

To ensure that the cermet does not react internally during service the combination of matrix and ceramic has to be chosen carefully. Acceptable pairs are UO_2 in aluminium or UO_2 in stainless steel, but UO_2 dispersed in zirconium and uranium carbide in beryllium cannot be used because they react either at the working temperature of the reactor or during fabrication. A dispersion of uranium or uranium alloy in stainless steel is unattractive for fuel purposes because of the formation of low-melting-point constituents and brittle intermetallic compounds.

The dispersion of a fissile ceramic such as UO_2 is satisfactory up to 1,400°C. and is achieved by powder methods. The elemental powders are mixed, pressed, and sintered in hydrogen. The sintered mass is then worked by an operation such as coining and clad in wrought stainless steel by co-rolling, co-extrusion or swaging. Austenitic stainless steels appear to be preferred over ferritic types because of superior corrosion resistance, especially to pressurized water systems. If there is more than 40% of dioxide in the UO_2/stainless-steel system, difficulties are experienced with fabrication and subsequently under irradiation. Reactor designs using highly enriched uranium require a dispersion of about 15–30 vol.% of UO_2. A fully austenitic stainless steel of the 25/20 Ni/Cr type has neutron absorption cross-section of about 2·5–3·0 barns. If a higher cross-section of 3·0–3·5 barns is acceptable in an enriched fuel reactor it is possible to disperse UO_2 in the higher nickel-chromium and nickel-chromium-iron alloys and take advantage of their superior high-temperature properties. Using a technique of cold pressing, sintering in hydrogen at about 1,150°C. followed by hot-rolling, almost theoretical density is obtained in these alloys and their high-temperature properties are about 90% of those of the cast and wrought version. It is necessary to ensure that the UO_2 particles are regular in shape, strong and dense, or the hot-working of the stiff matrix breaks up the particles and leads to local segregation in the form of stringers.

The reaction between UO_2 and zirconium is negligible below 550°C., but is significant at 700°C. and very rapid at 1,100°C. The uranium dioxide appears to be decomposed at high temperatures into its component elements which diffuse readily into the zirconium. The reaction would not prevent a UO_2-zirconium cermet fuel being used in a reactor at temperatures up to 550°C., but the reaction certainly imposes severe limits on the fabrication of the cermet.

Beryllium is attractive as a matrix for thermal reactor systems, but for the two commonest ceramics, UO_2 and UC, its use is limited to temperatures below 600°C. Above this temperature a solid-state reaction occurs between beryllium and either the dioxide or monocarbide of uranium and leads to an increase in volume in the neighbourhood of the reactants. The creep strength of beryllium also drops sharply above 600°C., so it would not be expected to retain fission gases very effectively.

Generally speaking, if reactor maximum fuel temperatures of greater than 600°C. are required, none of the metals with low neutron-capture cross-sections can be considered as cermet matrices and attention must be given to matrices based on iron, nickel or molybdenum. *to be continued*

Vacuum brazing furnace

An order for a high-vacuum brazing furnace worth more than £7,500 has been placed with Vacuum Research (Cambridge) Ltd., by the Admiralty. This will be an electric-resistance furnace to be evacuated to less than 10^{-4} mm. mercury. A temperature of 1,300°C. can be maintained in a hot zone 2 ft. long by 1 ft. square. A shielded window is provided to enable the operator to view specimens within the furnace.

TABLE 16 *Properties of some possible fissile dispersions*

Compound	Density	Fissionable vol. ratio	Melting point, °C.	Thermal neutron-capture cross-section, barns	Thermal conductivity, c.g.s. units	Structure
U	18·9	1·0	1,120	—		
U_3Si	15·6	0·77	930	0·13	0·036	Tetragonal
U_3Si_2	12·2	0·59	1,665	0·13	0·035	"
UN	14·3	0·71	2,630	1·88		
UC	13·6	0·69	2,270	0·003	0·080	NaCl cubic
UC_2	11·7	0·56	2,400	0·003		B.C. tetragonal
UO_2	10·96	0·53	2,500	0·002	0·014	CaF_2 cubic
UAl_4	6·0	0·22	730	0·23		
UAl_3	6·7	0·26	1,320	0·23		
UAl_2	8·9	0·35	1,500	0·23		
US	10·9	0·51	2,000			
PuO_2	11·5	0·53	3,200			CaF_2 cubic
PuC	13·9	0·70				NaCl cubic

NEWS

Leipzig Spring Fair

Colvilles of Glasgow, Britain's only makers of stainless-steel clad plates, are to exhibit for the first time at the Leipzig Spring Fair on March 5. A company spokesman said today Colvilles were already supplying large amounts of this product to the German Democratic Republic (East Germany).

The plates, marketed under the trade name 'Colclad,' form a stainless-steel veneer on ordinary steel plates, giving corrosion resistance of particular use in chemical and oil installations.

About 200 British firms will be represented at the Leipzig Fair which lasts for 10 days. In the steel section alone, British firms have increased from 35 last year to 51 in 1961, and floor space in this section has risen from about 6,260 sq. ft. in 1960 to 12,420 sq. ft. in 1961.

Powder metallurgy

The Powder Metallurgy Joint Group of the Iron and Steel Institute and the Institute of Metals announces that the following meetings will be held in 1961:

Discussion on 'The appraisal of powders for pressing and sintering: 1.—Techniques for the evaluation of powders. 2.—The relationship between powders and their pressing and sintering behaviour,' Monday and Tuesday, April 17 and 18, at the Royal Commonwealth Society, Craven Street (near Northumberland Avenue), London, W.1.

Symposium on 'Sintered high-temperature oxidation-resistant materials,' Thursday and Friday, December 7 and 8, at Church House, Great Smith Street, London, S.W.1.

Offers of original papers for this symposium should be sent to the Secretary, Powder Metallurgy Joint Group, 17 Belgrave Square, London, S.W.1.

BISRA 'open days'

The British Iron and Steel Research Association will hold two 'open days' at its Sheffield Laboratories on Thursday and Friday, June 15 and 16. The last occasion when the Hoyle Street group of laboratories held 'open house' for representatives of member firms was in 1956. This year's function will, therefore, give visitors the opportunity of examining the many developments which have taken place since then.

The Hoyle Street premises house three of BISRA's five main divisions—Steelmaking, Metallurgy (General) and the Mechanical Working Division. In addition, the Steel User Section, which operates an important advisory service to industry generally, is also housed there. Among the many displays and demonstrations planned by Steelmaking Division will be the latest developments in continuous casting, and work on spray refining and rapid desulphurization. One of the Mechanical Working Division's projects on show will be a recently-developed rapid-annealing system. Prominent among the Metallurgy Division's projects will be aspects of its general preoccupation with the purification of steel (jet de-gassing, vacuum casting, etc.), work on improved speed and accuracy in analysis, and the production of high-temperature steels. All these examples are arbitrarily selected, and represent only a few of the many projects which will be on show.

One-day instrumentation conference

The Society of Instrument Technology, and the Birmingham Productivity Association, are holding a one-day conference, at the Birmingham College of Technology, Gosta Green, Birmingham, on Wednesday, March 29.

The conference is aimed at the small- to medium-sized manufacturer, and the object is to show the various aspects of instrumentation which will increase productivity profitably.

An Instrument Exhibition has been arranged with the Conference, at which instruments which are discussed during the day are exhibited. Potential users will be able to discuss their ideas with instrument experts, and the necessary contacts made for future use.

Kepston furnaces

British Furnaces Ltd., of Chesterfield, in association with Kepston Ltd., of Kinross, have formed a Kepston Division of British Furnaces, for the production of electric-resistance furnaces, to operate with controlled atmospheres, including hydrogen, nitrogen, argon, and helium, for heat treatment and brazing operations of most metals and stainless steels. The outstanding features of the Kepston furnace include the complete absence of normal hydrogen hazards, controlled heating rates from very slow, up to 100°C./min., to 1,600°C., controlled cooling from vacuum slow to 100°C./min., down to 450°C., no restriction on size and shape of heating zone from laboratory size upwards, and operations under visual control at all temperatures.

All electric-resistance furnace enquiries will be handled by the Kepston Division of British Furnaces Ltd., Derby Road, Chesterfield.

LARGE, MAN-MADE DIAMONDS

A group of carat-sized diamonds made by scientists of the General Electric Research Laboratory in Schenectady, N.Y., seen against a 1-in. scale. These large diamonds are dark in colour and are not yet of sufficient mechanical strength for industrial application. General Electric can now make, however, in the laboratory, diamonds up to $\frac{1}{16}$ of a carat in size that are of excellent industrial quality. Stones of this size are needed for metal-bonded diamond wheels and saws. These $\frac{1}{16}$ of a carat diamonds are not yet available commercially. Industry also uses diamonds in full carat ranges for drills, dressing tools, dies, and single point cutting tools. 'If General Electric can perfect the mechanical structure and improve the strength of its new, carat-size diamonds, the company will be able to compete with natural stones in the full industrial range,' according to John D. Kennedy, general manager of G.E.'s Diamond Production Business Section in Detroit

NEW PLANT

Instrumentation for carbide sintering

Following a trial period with an experimental installation, Associated Electrical Industries (Rugby) Ltd. are incorporating an advanced form of automatic temperature control on electric resistance sintering furnaces at their Ardoloy factory, Rugby.

Conversion of four furnaces to automatic working is now nearing completion and a total of eight automatic installations are expected to be in operation by the end of this year. The technique adopted is based on the Electr-O-Volt proportional control system developed by Honeywell Controls Ltd. and fitted integral with their ElectroniK controller instruments.

The main advantage derived is that the working temperature is now maintained within 2°C. automatically as compared with 20°C. using the previous manual control. There is found to be a rapid response to load fluctuation without undershoot or overshoot. Sintering of Ardoloy cemented carbides is carried out at temperatures between 1,350°C. and 1,500°C., and the output per furnace can vary between 30 and 150 Kg. of tips per week depending on the grades, size and shape of the pieces being sintered. Thus, in addition to the response to load fluctuations, a very desirable feature in a control system is rapid change from one working temperature to another. This is, in fact, obtained by the new automatic scheme in a straightforward and simple manner. Elimination of all mechanical moving parts outside the control instrument contributes much to the exactness and high response speed achieved. The original manual technique involved the use of hand wheels. As an interim measure, electric motors were fitted to these wheels but they took some 30 sec. to complete the full travel. The new scheme, in contrast, corrects changes within a few milliseconds.

Dunking machine

Pneumatic operation of the entire dunking action, both upwards and downwards, is an important new development incorporated in the heavy-duty dunking machine recently introduced into the Dawson range of metal-cleaning and degreasing machines.

Manufactured by Dawson Bros. Ltd., of Gomersal, near Leeds, this dunking machine is intended for applications where soaking and agitation are more suitable than other metal-cleaning methods. Typical examples of its variety of uses are the washing of sand from castings and the rough cleaning of cylinder blocks.

The dunking action creates a strong turbulence in the cleaning solution which scours all the surfaces of the parts and penetrates recesses, loosening and removing the sand, grease and other residues.

Operation is simple ; the parts to be cleaned are transferred from the feed roller conveyor on to the dunking table which the operator lowers into the cleaning soluton by means of a control valve. When the table reaches the bottom, the dunking starts immediately. At the end of the cleaning period, the operator reverses the control val e which stops the dunking action and elevates the table. The next batch of soiled parts is pushed on to the table at the same time ejecting the cleaned parts on to roller conveyor at the other side of the machine.

A filtration system can be fitted to the machine to provide continuous filtration for the cleaning solution.

This machine has been manufactured as a dual dunking unit for washing and rinsing and is so designed as to enable any number of units and alternative processes to be linked in-line. This system could be arranged for fully-automatic operation with a moving beam or similar system fully timed and interlocked. A particularly suitable application would be in the field of phosphate coating. Dawson Bros. Ltd., Gomersal, near Leeds. Sole agents: Drummond-Asquith Ltd., King Edward House, New Street, Birmingham 2.

Induction gear hardening

A new universal induction gear hardening machine has been designed by Induction Heating Equipment Ltd. for a leading firm of motor-car manufacturers principally for the hardening of gears. The power source is an inductor alternator working at 500 V. 10 kc./s. The ancillary equipment consists of a control cubicle and two work stations which are used alternately. The switching is so interlinked that power can only be switched to one station at a time although both may be loaded. Either a hand-operated or automatic cycle can be used, and provision is made for loading work either above or below the inductor.

Each station has a separate transformer fitted close to the work; to the secondary of which the inductor is bolted. The inductor is therefore stationary, and both the work and quench tank move up and down independently of each other. With the controlling switch set in the ' hand ' position, heat, quench, rotation and work movements may be controlled at will by separate switches. In the ' automatic ' position, two conditions are available:

In the first case the loading position is presented above the inductor and pressing the ' start ' button causes the work to commence rotating and move downwards into the heating inductor, and power is switched on automatically if the other station is not taking power. If this should be so, the work remains in the coil until the other station has finished its heating cycle, when heat comes on automatically. The heating cycle is timed by synchronous timer, and on completion the work will lower itself into the quench tank which has in the meantime been raised to a position immediately below the inductor. After a pre-

1 *Induction gear-hardening machine*

set time interval the work is returned to the loading position for removal, and a fresh work-piece is inserted.

Certain shapes of work-piece, such as cluster gears, cannot be passed completely through the inductor unless the inductor is made so large as to be inefficient. By changing over the selector switch, the loading position is presented below the inductor, the quench tank being lowered out of the way. Pressing the 'start' button causes the work-piece to move upwards into the inductor; the quench tank also moving upwards. After a preset heating period the work is lowered rapidly into the tank, and the tank subsequently lowers, exposing the loading position for removal of work.

Oil or water may be used as the quench medium, and this is kept cool by recirculation through an external heat exchanger, the level in the quench tank being maintained by weir action. The liquid is pumped into the tank through a semi-circular pipe surrounding the work and drilled on its inner side so that a swirling motion is maintained around the work-piece, thus cooling it more effectively.

Alcan furniture alloy

A new extrusion alloy developed by the Aluminium Co. of Canada Ltd. expressly for use in furniture manufacturing has produced highly satisfactory results.

The new alloy, called Tube-Alloy, makes it possible to extrude aluminium tubing easily and economically in an almost limitless variety of shapes—square, triangular, octagonal and asymmetrical designs. In addition the tubing can be heat treated to strength levels equalling those of extruded and drawn tubing in other aluminium furniture tubing alloys. This development puts at the disposal of furniture manufacturers an alloy which is not only economical but provides designers with much more latitude in achieving style and sales appeal.

Alcan's Tube-Alloy is a member of the magnesium silicide group of aluminium alloys. It is heat-treatable and develops higher strength than extruded 50S, the normally-used alloy in the furniture-making industry. It is also less expensive. Although it can be cold drawn, it is preferable to extrude this alloy to the required size, thus eliminating one operation.

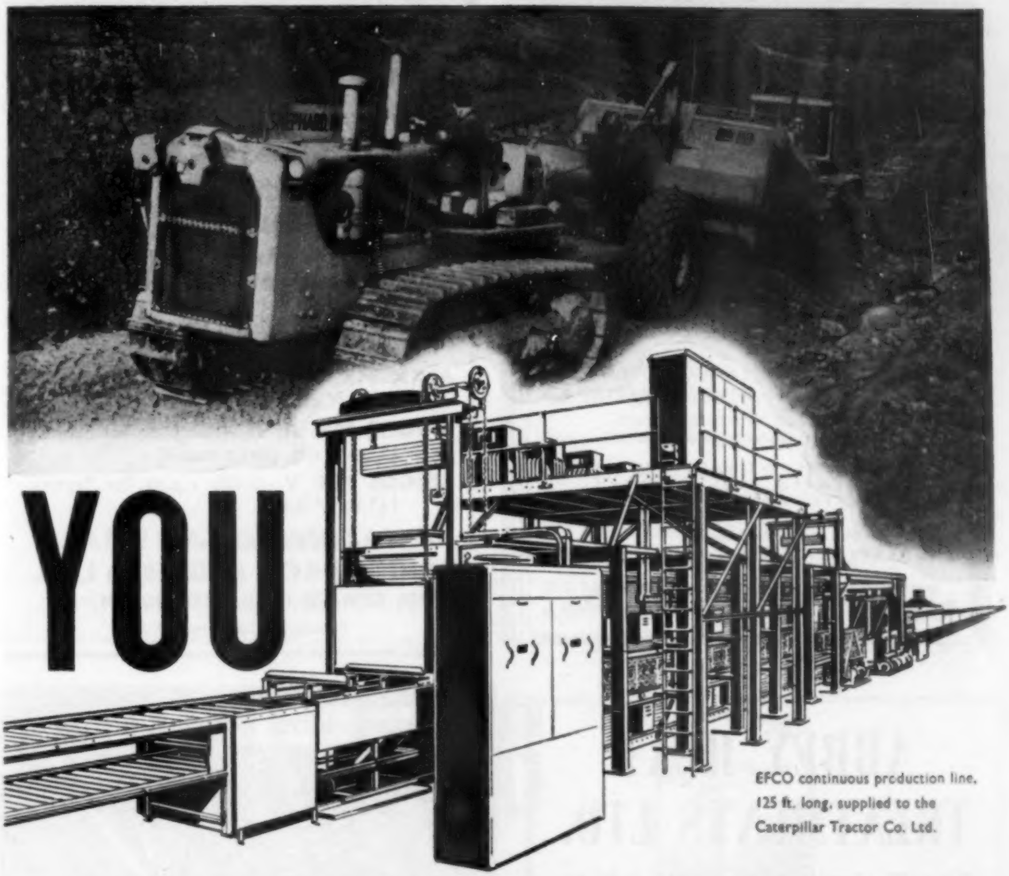
When extruded in accordance with recommended practice, it takes on a fine-grained structure which permits bending without the formation of nodules known as 'orange peel.' Normal polishing and buffing give it a good finish.

Developed expressly for the furniture industry this new alloy has not only performed extremely well in this field but gives promise of being equally applicable in other areas where high fatigue and impact properties are not required.

Basin tilter furnace

The Morgan Crucible Co. Ltd. has now extended its range of crucible furnaces by producing a furnace, the B.T.1300, of 1,300 lb. aluminium capacity. It is designed on similar lines to the two very successful smaller furnaces known as the basin tilter types B.T.380 and B.T.500. The range of furnaces is designed to provide simple and flexible melting units embodying all the features stipulated by the aluminium die-casting industry. (The two smaller furnaces are also being extensively used for the melting of copper base alloys.)

The B.T.1300, considering its large capacity, is a very compact unit; the low platform and large diameter crucible facilitate the charging of bulky scrap. Although particularly suitable as a bulk melter for the die-casting foundry supplying metal for 'bale-out' furnaces, it is equally successful when used on its own, giving many important advantages.



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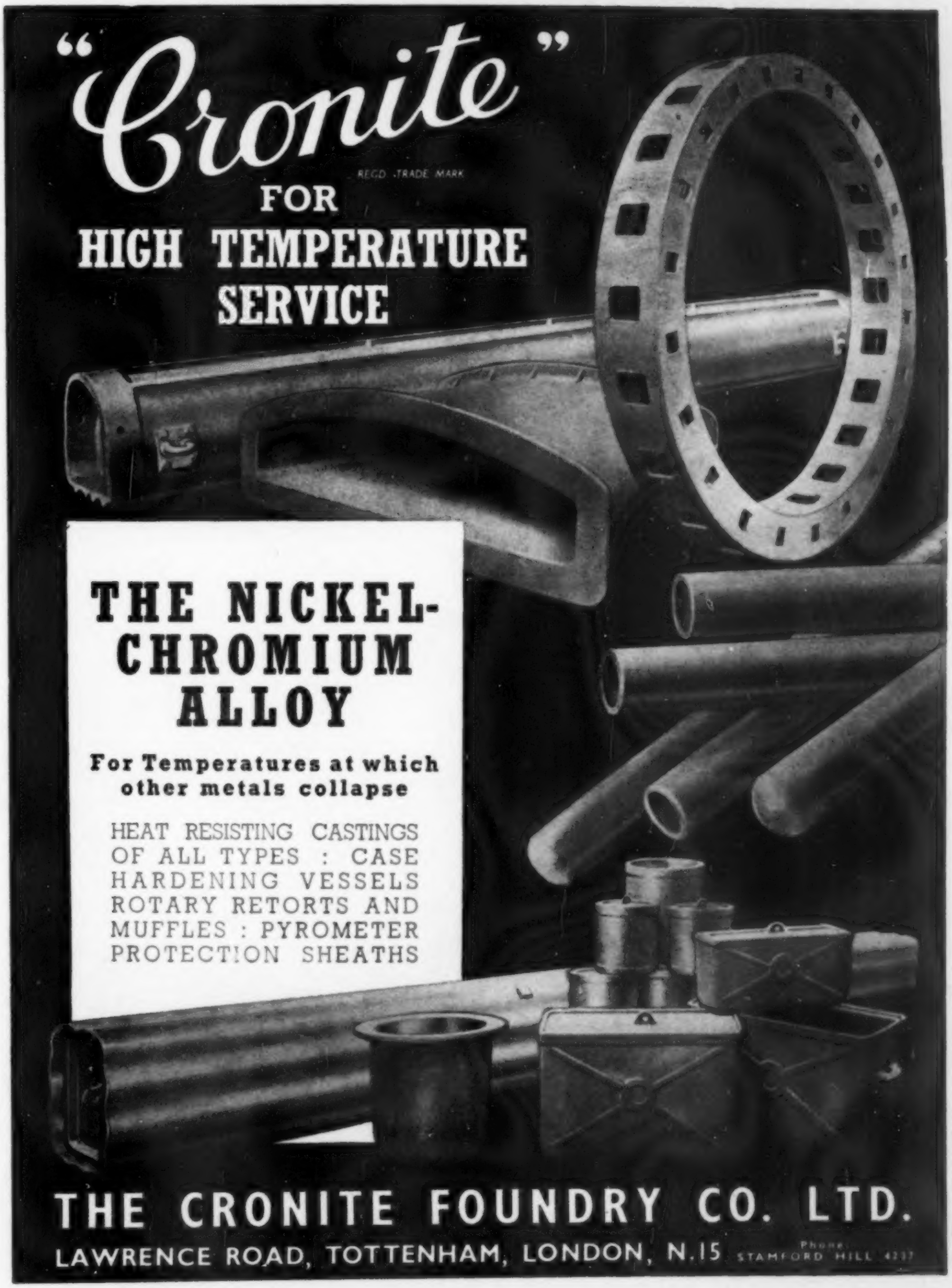
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